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DONALD ARTHUR OLIVER

DONALD ARTHUR OLIVER was born in February, 1906. He graduated with First Class Honours (University of London) in Electrical and Mechanical Engineering with additional Mathematics and Metallurgy, and was awarded the degree of M.Sc. for two researches : the first on new criteria of ductility for steels, and the second on the velocity of sound in sheet materials.

After two years at the William Froude National Tank at the National Physical Laboratory, he transferred to the Electrical Measurements and Standards Department, under the late Dr. D. W. Dye, F.R.S., and dealt with dielectrics and magnetic materials. During this period he assisted Dr. L. Hartshorn with developments in alternating current bridges, and collaborated to redetermine the dielectric constant of pure benzene to high accuracy.

Under the late Sir Clifford Paterson, F.R.S., at the Research Laboratories of the General Electric Co., Ltd., Wembley, he was responsible for setting up a new electro-acoustics laboratory, and evolved a new type of condenser microphone now used as a substandard in the Physics Division, N.P.L. He also studied the design of permanent magnet systems.

In 1934, Mr. Oliver was appointed to the staff of William Jessop and Sons, Ltd., Sheffield, first in charge of permanent magnet design, and four years later as first Director of Research. From 1938 onwards he served on a number of the Joint Committees under the British Iron and Steel Federation and The Iron and Steel Institute, including the Steel Castings Committee, the Ingot Moulds Sub-Committee, the Hair-line Crack Sub-Committee, the Inclusions Sub-Committee and the Liquid Steel Temperature Sub-Committee. He was also Chairman of the Foundry Steel Temperature Sub-Committee. He was a member of the Technical Advisory Committee to the Alloy and Special Steels Committee of the Ministry of Supply, and Chairman of its Hardenability Sub-Committee. He was a member of the British Metallurgical Mission (1943) to the United States and Canada, dealing with ferro-alloy conservation, and was a member of the British Steel Mission to Germany (1945). He was a member of the Metallurgical Committee of the Ministry of Supply from 1941 to 1944. He is also a member of several British Electrical and Allied Industries Research Association Committees and is Chairman of Committee N/C on Permanent Magnets. In 1944 he was appointed Director of Research to the Birmingham Small Arms Group of Companies, of which William Jessop and Sons, Ltd., is a member firm.

In the British Iron and Steel Research Association, Mr. Oliver has served as a member of Council and on the Panels dealing with Steel Making, Steel Castings and Metallurgy (General). He is a member of the Alloy Steels Research Committee. He is Chairman of the Metal Physics Committee of B.I.S.R.A. and is also a member of the Metal Physics Committee of the Institute of Metals. He is a member of the Inter-Service Metallurgical Research Council and of the Heat Resistant Materials Committee.

Mr. Oliver is a Fellow of the Institution of Metallurgists, a Vice-President of the Institute of Physics and President of the Society of Instrument Technology. He became a Member of The Iron and Steel Institute in 1940, and was elected a Member of Council in 1947.

Vandyk

Donald Arthur Oliver M.Sc.
Member of Council

IRON AND STEEL
INSTITUTE PAPERS

PLANT DESIGN · MANUFACTURE · RESEARCH · DEVELOPMENT

The Effect of Sodium Oxide Additions to Steelmaking Slags

Part I—Use of Soda to Dephosphorize Pig Iron at 1400° C.

By W. R. Maddocks, Ph.D., and E. T. Turkdogan, M.Met.

SYNOPSIS

Soda has been added to basic and acid slags with the object of bringing about dephosphorization at low temperatures. In basic slags it has been found impossible to retain soda as a silicate which is necessary for the formation of sodium phosphate. With slags containing up to 50% silica and a low CaO/SiO_2 ratio, dephosphorization has been effected after melting on acid hearths. Under favourable conditions as much as 85% dephosphorization of a 1·34% phosphorus iron has been brought about.

Introduction

During the present century an extensive literature has been published concerning the removal of phosphorus from iron, and great prominence has been given to the physical chemistry of the dephosphorization reactions. It is now well established that the removal of phosphorus from the metal takes place by oxidation and by the formation of more or less stable phosphates which are taken up by the slag. Calcium phosphate is by far the most common form in which the phosphoric acid occurs in basic slags, since it is the most stable of all the possible phosphates that have been found practicable up to the present, at the steelmaking temperatures. The process of dephosphorization under lime-bearing slags has been defined by well-established equilibrium conditions. One of these conditions indicates that, other things being equal, the removal of phosphorus is greater at lower temperatures. There would be obvious advantages, if it were found possible to remove phosphorus from pig iron at temperatures as low as 1400° C.

Up to the present, there have been practical difficulties in producing slags containing sufficient lime, which are fluid enough at such low temperatures to allow efficient dephosphorization. If a suitable base, other than lime, were used which would produce stable phosphates, act as a flux, and at the same time be capable of solution in a slag in such a manner that phosphates could form, most of the difficulties met with hitherto would be overcome.

Sodium oxide appeared to be the most suitable base for this purpose. It was known to form a very stable phosphate and silicates that were of low melting temperature. The information available on the use of soda was not promising, and since it behaves as a strong base the obvious approach seemed to be its introduction into basic slags. It required a relatively small number of experimental melts to show that this approach was fundamentally wrong, and that if a reserve of soda were to be made available for dephosphorization, acid slags containing little or no lime would have to be employed.

PROCEDURE

EXPERIMENTAL PLANT

The electric-arc furnace used for the experiments was originally designed and built by previous workers, in whose paper[1] a detailed explanation of the plant and the experimental technique was given. Here, only a short description of the plant will be made; for its detailed development reference should be made to the previous paper.

A vertical section, and plan sections of the electric-arc furnace, which consisted of three layers of chrome-magnesite bricks, cut to shape, and held together in steel clamps, are shown in Figs. 1 and 2. The bottom chrome-magnesite brick had a hearth which was dish

Manuscript received 27th August, 1948.

Dr. Maddocks is Senior Lecturer in Ferrous Metallurgy, and Mr. Turkdogan a research student, at Sheffield University.

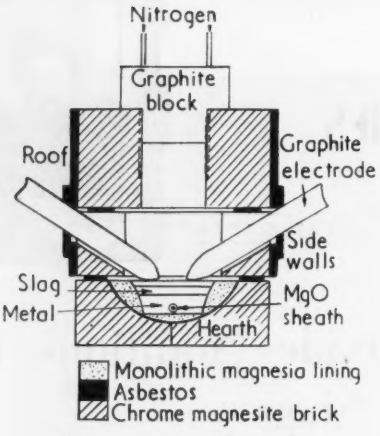


Fig. 1—Section of furnace

shaped, approximately 6 in. dia. at the top, with a depth of 2 in.

The melts were made on two different types of hearth, both of which were dried in air for two days before use :

Magnesia Hearth—250-mesh electrically fused magnesia was bonded with a dilute solution of sulphite lye, and this mixture was rammed into the space between an iron former and the chrome-magnesite brick. The hearth was circular in horizontal cross-section, and tapered from top to bottom, with dimensions of 4 in. at the top and 2¾ in. at the bottom.

Silica Hearth—200-mesh silica flour was used in exactly the same way as described for magnesia.

A magnesia thermocouple sheath was used for the magnesia hearths, and a silica thermocouple sheath for the silica hearths. Having made the hearths, the channels were cut in the walls so that thermocouple sheaths could be inserted from opposite sides, and the gaps were rammed with the mixture that was used to make the hearth. The position of the sheath was so adjusted that it was about ¼ in. below the slag–metal surface.

Little difficulty was experienced with the thermocouple sheaths in regard to their failure. The occasional failure of the thermocouple, in the present research, was always accompanied with the breakdown of the sheath, and that occurred if the sheath was very near to the slag–metal layer.

Temperature measurements were made with a platinum–platinum/13% rhodium thermocouple connected to the Tinsley D.C. amplifier, which was fitted with a recorder and a large dial indicator.

The A.C. current from a 500-V. main passed through a Ferranti 110-V., 16·5-kVA., oil-cooled welding transformer, and then through fixed, and variable, oil-cooled chokes. The electrodes were connected to the variable oil-cooled choke so that the current through the electrodes could be changed by moving the core of the choke. The variation in the current density by means of this choke was a most convenient method of controlling the temperature of the bath. The electrical circuit is shown in Fig. 3.

The slag and metal samplers were of the same type as used by the previous workers. The slag sampler consisted of a copper block 2 in. dia., and 1 in. thick, which had grooves near the lower edge and across a diameter of the lower surface in order to hold the slag after solidification.

The metal sampler was similar to that used by Taylor and Chipman.[2] It consisted of a 3½-in. length of 1-in. dia. copper rod which was reamed with a standard No. 6 tapered-pin reamer. A male ½-in. gas thread was machined at the upper end of the copper rod, and the connection to the aspirator bulb was made with standard gas fittings.

The slag sampler was degreased before taking a slag sample, otherwise the slag did not adhere to the sampler. On the other hand, the copper portion of the metal sampler was always greased with vaseline before use to prevent the slag and the metal sticking on to the copper.

PIG IRON AND SYNTHETIC SLAGS

All the experiments were carried out with high carbon and high phosphorus pig-iron charges. Because of the difficulty in obtaining a pig iron of the required composition, the pig iron used in the experiments was made by melting a Swedish white pig iron with an iron–phosphorus alloy containing 25% phosphorus. Sixty pounds of this metal were melted in a crucible in a gas-fired furnace, and cast into sand moulds. These small pigs were about 1 in. deep, ½ in. thick, and 1 ft. long. Before charging the furnace these pigs were thoroughly cleaned by filing and brushing, and broken to a size of about 1 × 1 × ½ in.

The pig iron used throughout the experiments was a low silicon basic pig iron which had the following composition :

C, %	P, %	Mn, %	Si, %	S, %
2·94	1·34	0·32	0·14	0·05

For the preparation of the synthetic slags the following commercially pure chemicals were used :

(i) Precipitated silica, containing about 12% of moisture
(ii) Precipitated chalk, which was calcined immediately before use
(iii) Tri-basic sodium phosphate
(iv) Sodium carbonate
(v) Ferric oxide
(vi) Manganous carbonate.

The required quantities of the necessary chemicals were weighed out, and mixed in a small ball-mill for about an hour. Since the mixture, in its powder form, was very unsuitable and inconvenient for handling during addition to the furnace, it was fused in two ways ; with and without sodium carbonate.

(1) *Mixtures without Sodium Carbonate*—A portion of the mixture was transferred to a silica

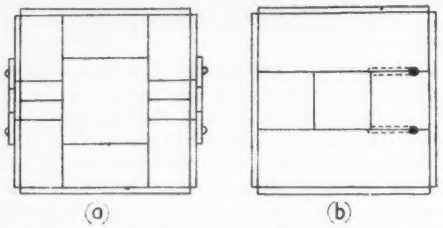


Fig. 2—Plans for (*a*) furnace side-walls and (*b*) roof

evaporating dish, and heated up by an electric arc struck between ½-in. dia. carbon electrodes carrying a current of 20–35 amp. By the adjustment of the electrodes and the core of the variable choke, the fusion of the mixture could be controlled. The compact lump of the fused slag was made so that it weighed about 25 g.

This method of preparing the synthetic slags was very sound for the following reasons :

(*a*) During the fusion operation there was no danger of contamination

(*b*) Having mixed the chemicals well in a steel ball-mill, one could control the composition of the slag within close limits, provided gas-evolving compounds were not abundant. Otherwise, the evolution of an excessive amount of carbon dioxide, or similar gases, during the ignition of the compounds would cause the powdered chemicals to be scattered, particularly the lighter compounds such as silica. For this reason pure ferric oxide was preferred to ferrous oxalate

(*c*) The high temperatures reached by the arc made it possible to obtain homogeneous synthetic lumps.

(2) *Mixtures with Sodium Carbonate*—The fusion of these mixtures in an electric arc was not very successful because there was an enormous loss in soda during melting before it formed any ferrites or silicates ; but sintering was found to be most satisfactory. The mixture was transferred to a 5-in. dia. fused-silica, or porcelain dish, and heated up slowly in a gas-fired muffle till it formed a clinker. Slow heating was essential to allow gentle evolution of the carbon dioxide, and, as the temperature rose, soda (in future the term soda will be used for sodium oxide, Na_2O) combined with ferric oxide and silica to form ferrites and silicates. The melting points of these ternary compounds being somewhere between 850–950° C., special care was taken not to fuse them, since there was the risk of contamination by other materials from the fused-silica or porcelain dish. When the mixture became sticky and soft the gas was shut off, and the contents of the dish were allowed to cool in the muffle. When it was cold the clinker was easily removed from the dish.

The amount of soda lost during the sintering operation could be found, at least approximately, by checking up the weights before and after the ignition. Generally speaking, the loss in soda was high in those mixtures containing lesser amounts of ferric oxide and silica.

EXPERIMENTAL PROCEDURE

Only small alterations were made to the experimental technique of previous workers.[1] As mentioned before, the pig iron was first cleaned and broken into small pieces about 1 × 1 × ½-in. size. After drying the hearth, whether magnesia or silica, in air for two days, its surface was thoroughly cleaned with a soft brush, and 600 g. of pig iron were charged. The side walls of the furnace were then placed in position, and the electrode holders were positioned so that all necessary adjustments could be made during the melting operation. Finally, the roof and graphite lid of the furnace were placed in position. To ensure a good contact between the bricks, thick asbestos cords were packed between each clamp of bricks.

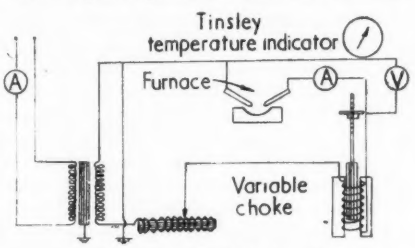


Fig. 3—Electrical circuit

The melts were made in an atmosphere of nitrogen to prevent the oxidation of iron, and to ensure the control of the iron content of the slag. After the nitrogen had passed through two openings in the top of the roof of the furnace for some time the current was switched on. During the first 10 min. the furnace wall and roof were warmed up by keeping an indirect arc between the electrodes which were inclined about 20–25° to the horizontal. During the following 10-min. period the current was gradually increased from 100 to 150 amp. by raising the core of the choke, and, as the metal started to melt, the degree of inclination of the electrodes was increased.

Because of the high carbon and high phosphorus content of the pig iron, the charge became completely molten in about 20 min. from the start. When this stage was reached the synthetic slag clinker was added in small amounts at a time in order to prevent chilling during the slag-forming period. The slag weight was about 200 g., and the time taken for making the complete addition was about 20 min. By this time, the slag and metal were completely molten, and the bath was quite ready for oxidation. At this stage a platinum–platinum/13% rhodium thermocouple was inserted in one of the sheaths.

Now, the angle made by the electrodes with the horizontal was 40° in order to obtain a direct arc with a distance of about 1 in. between the tips of the electrodes. This direct arc gave a uniform temperature. During the refining period the temperature varied between 1390 and 1410° C. At temperatures above 1420° C. the rate of ' boil ' rapidly increased.

After the final addition of the synthetic slag a period of 20 min. was given for the bath to attain an equilibrium condition, and, as a rule, at the end of this period, the boil had reduced considerably, and in most cases the bath was quite still at the time of sampling. During this 20-min. period of equilibrium great attention was paid to the constancy of the temperature, which was required to be at 1400° C. It was not difficult, however, to control the temperature between the maximum limits of ± 10° C.

Spare thermocouples also were available in case of any failure of the thermocouple in action. As mentioned earlier, the failure of the thermocouple often coincided with the breakdown of the magnesia or silica sheath, which was most likely caused by the penetration of slag through the sheath, which happened more frequently if the sheath were positioned near the slag–metal surface. Furthermore, to prevent the failure of the thermocouple and to ensure freedom from contamination during the melt, the hot junction was renewed once or twice.

Having kept the bath at a constant temperature for the required length of time, the arc was switched off, the electrodes and thermocouple were withdrawn, and the metal and slag samples were taken through the hole exposed by removing the graphite lid.

Metal Sampling

The aspirator bulb was depressed and the sampler pushed through the slag layer into the underlying metal; then the bulb was released, and the sampler withdrawn almost immediately, and quenched in clean cold water. Although the carbon content of the refined metal was very high, decarburization during the course of sampling was slight.

Slag Sampling

A sample of slag was obtained by immersing the copper block portion of the sampler into the slag layer and quenching the slag adhering in cold water. After the slag and metal samples had been taken, the roof of the furnace was removed and the slag was allowed to cool in air.

The attack of the soda-containing slags on the magnesia and silica linings was not excessive at the temperature employed. No alumina nor magnesia was added to the synthetic slags, and therefore the amounts of these constituents in the final slag may be taken as a measure of the attack of the slag on the lining.

Although both types of hearth were unfired previous to the melt, *i.e.*, the hearths were in a green state, the slag penetrated into the lining of the hearth to a depth of less than $\frac{1}{4}$ in. It was possible, however, to preheat the hearth, and, in fact, a glazed surface had been obtained on a silica hearth by heating up the furnace by an indirect arc; but the surface of the lining was not smooth and minute surface cracks became somewhat larger. In addition, the replacement of the thermocouple sheath into its position, after glazing the surface, offered some difficulty. On the other hand, prefiring shortened the total time of the melt by about 20 min.

The total time taken for a heat from start to finish varied from 1 hr. to 1 hr. 20 min. Shorter total times were preferred because no undue oxidation of carbon took place, and also because there was less attack on the linings. Both conditions were obtained by heating up the furnace quickly during the melting-down period.

METHODS OF ANALYSIS

Chemical Analysis of the Metal

The metal samples, being high in carbon and quenched in water, were very hard and brittle, and drilling or machining of these samples was impossible. Consequently, when the surface of the pencil sample was cleaned by means of a file, it was crushed in an iron mortar, and the powder obtained was very suitable for analysis.

The refined iron was analysed for carbon and phosphorus only. The carbon was determined by the combustion method, and the phosphorus was determined by the gravimetric lead molybdate method.

Chemical Analysis of the Slag

Since the slags contained soda, *i.e.*, Na_2O, there was a great possibility that some of the constituents would dissolve on water quenching, and, therefore, air-cooled slag samples were preferred for analytical purposes. The question of oxidation of ferrous iron during the air cooling of the slag was considered, but the oxidation of ferrous to ferric iron was not thought to take place to any great extent, for the following reasons:

(i) The rate of cooling of the slag in air was high enough to suppress this reaction
(ii) The high silica content of these slags rendered the oxidation reaction very sluggish.

In fact, the analysis of water-quenched and air-cooled slags for iron did not show any oxidation at all.

The slag sample was prepared for analysis in a conventional way. It was crushed with a percussion mortar to pass a 30-mesh sieve, and any particles of iron present were removed by means of a strong magnet. To ensure the homogeneity of the sample, the 30-mesh slag was mixed in a clean small ball-mill for about half an hour, and a suitable portion of the slag was obtained from the crushed and mixed product by taking samples at nine points. This was further crushed to pass a 90-mesh sieve, and finally ground to a fine powder in an agate mortar for analysis. Before weighing out the sample for analysis it was dried in an oven at 110° C.

Dissolution of the Slags Obtained in the Experiments

Those slags containing less than 45% silica were soluble in dilute 1 : 1 HCl solution, but those containing more than 45% silica had to be fused in sodium carbonate for easy decomposition. No attempt was made to find the degree of solubility of the soda-slags in various solutions, but it had been observed that as the amount of soda decreased, the solubility of the slags in dilute 1 : 1 HCl solution also decreased. In the absence of soda the slags containing more than 30% of silica had to be fused with sodium carbonate in order to bring about solution.

Each slag was analysed for FeO; Fe_2O_3; P_2O_5; SiO_2; Na_2O; CaO; MnO; MgO; and Al_2O_3. The methods used for the analysis of each constituent are mentioned below:

ΣFe—Stannous chloride reduction method.

FeO—Decomposition of slag in a neutral atmosphere in a conical flask, fitted with a Bunsen valve, and titration with standard potassium dichromate solution, using 0·2% of barium diphenylamine sulphonate as an internal indicator.

Fe_2O_3—By difference of ΣFe and FeO.

SiO_2—Having decomposed the slag either by fusion or direct dissolution in dilute 1 : 1 HCl solution, it was taken to dryness on a sand bath, and silica was filtered off in the usual way. The filtrate was again taken to dryness to ensure that all of the silica had come out of solution.

P_2O_5—If the percentage of P_2O_5 in slag were higher than 2%, it was determined gravimetrically as $Mg_2P_2O_7$. If it were lower than 2%, the gravimetric lead molybdate method was used.

Al_2O_3—Having made the basic-acetate separation on the filtrate, from which silica had already been filtered off, the precipitate was dissolved in dilute 1 : 1 HCl solution, and alumina was precipitated

Table I
RESULTS OF CHEMICAL ANALYSES

Melt No.	Metal Analyses, Wt.-%		Slag Analyses, Wt.-%									Reduction, %	
	C	P	P_2O_5	SiO_2	CaO	Na_2O	ΣFe	FeO	Fe_2O_3	MnO	MgO	Al_2O_3	C	P
C1	0·97	0·68	7·11	26·80	46·76	8·37	3·53	3·00	1·72	0·40	4·34	1·50	67·01	49·25
C2	1·58	1·04	8·65	28·00	45·23	9·27	3·32	3·49	0·87	0·57	2·52	1·40	46·26	22·39
C3	1·51	0·95	9·25	26·22	45·17	10·05	4·33	3·93	1·82	0·62	1·91	1·03	48·64	29·11
C4	1·41	1·12	8·76	28·68	43·10	9·80	4·74	4·40	1·98	1·08	1·16	1·04	52·05	16·42
C5	1·25	0·71	9·15	25·87	42·87	10·86	4·99	4·67	2·23	0·73	1·62	2·00	57·49	47·01
C6	1·26	0·85	9·50	28·48	39·46	10·37	3·78	3·55	1·46	0·76	5·42	1·00	57·15	36·57
C7	1·52	1·23	13·50	24·44	36·64	12·57	2·29	2·30	0·72	0·71	6·67	2·45	48·31	8·21
C8	1·34	0·80	10·22	26·91	36·60	10·78	2·64	2·33	1·19	1·07	8·40	2·50	54·55	40·30
C9	1·43	0·41	11·30	20·18	30·60	9·47	4·45	4·01	1·90	0·76	19·21	2·57	51·37	69·40
C10	1·40	1·07	10·56	27·88	28·64	15·00	1·77	1·65	0·70	0·80	12·34	2·43	52·38	20·14

as $AlPO_4$. Double precipitation was applied to free the precipitate from iron contamination.

CaO and MgO—The solution, which was the filtrate from the basic-acetate separation, was used. Lime was determined by the ammonium oxalate double-precipitation method, and magnesia was precipitated as $Mg_2P_2O_7$.

MnO—The determination of MnO was carried out on a separate sample, using the bismuthate method.

Na_2O—Berzelius' method[3] was used for the determination of soda, which was as follows : Mix 0·5 g. of the slag in a platinum crucible with 5 c.c. of conc. H_2SO_4, and carefully add about 20 c.c. of pure HF. Heat gently on a sand bath, and when the HF has evaporated take the solution to dryness. Cool, add. conc. $NH_4.OH$; heat gently, and when the mass is quite disintegrated, filter and wash with water. Acidify the filtrate with HCl, and heat the solution to boiling, then add a slight excess of a hot solution of $BaCl_2$ to the boiling solution and let it stand for 12 hours.

Filter off $BaSO_4$ and wash with hot water ; then take the filtrate to dryness in a platinum dish, and ignite to drive off the ammonium salts. Dissolve the residue in 50 c.c. of water and 50 c.c. of pure absolute ethyl alcohol, and then add 50 c.c. of an alcoholic solution of ammonium carbonate, *e.g.*, Gooch and Eddy's solution, in order to precipitate the carbonates of barium, calcium, and magnesium. The precipitate, being bulky, absorbs some of the NaCl, so that, after filtering, the residue is dissolved and a second precipitation and filtration is applied. Evaporate the combined filtrates to dryness in a weighed platinum crucible, and add a few c.c. of conc. HCl and again evaporate to dryness. Ignite, cool, and weigh NaCl.

Then, weight of Na_2O = weight of NaCl × 0·5302.

This method was found to be very satisfactory, and in addition, reasonably quick. Although the quantitative analysis of soda was not difficult, it was very tedious and required special care. But, fortunately

Table II
RESULTS OF CHEMICAL ANALYSES

Melt No.	Metal Analyses Wt.-%		Slag Analyses, Wt.-%										Reduction, %	
	C	P	P_2O_5	SiO_2	CaO	Na_2O	ΣFe	FeO	Fe_2O_3	MnO	MgO	Al_2O_3	C	P
MA1	2·33	0·95	3·37	43·20	1·93	39·79	3·30	2·88	1·83	0·55	4·95	1·50	20·75	29·11
MA2	2·29	0·96	3·61	42·56	2·13	35·23	6·20	6·17	2·01	0·63	6·16	1·50	22·11	28·36
MA3	1·44	0·21	10·37	28·24	5·56	21·85	8·54	6·95	4·49	10·24	10·05	1·80	51·02	84·34
MA4	0·16	0·04	4·65	23·30	13·72	6·11	24·42	26·44	5·52	12·00	7·46	0·80	94·27	97·13
MA5	0·05	0·15	2·55	6·24	6·32	3·12	53·44	60·00	9·71	3·60	7·56	0·90	98·20	89·20
SA1	1·50	0·17	5·38	33·96	7·12	10·73	23·32	26·10	4·33	6·22	2·46	3·70	48·99	87·33
SA2	1·72	0·25	5·58	34·30	6·54	14·83	19·78	22·22	3·58	7·44	2·55	2·96	41·51	81·34
SA3	1·95	0·52	4·60	37·60	6·52	14·47	19·22	22·12	2·89	7·68	1·35	2·77	33·67	61·20
SA4	1·95	0·56	3·98	37·68	4·74	14·53	21·04	24·50	2·86	7·80	1·20	2·71	33·67	58·22
SA5	1·21	0·78	4·17	40·42	8·80	7·93	20·56	21·66	5·32	8·70	2·10	0·90	56·78	43·88
SA6	1·46	0·64	2·30	40·70	10·80	3·08	25·30	29·25	3·66	7·08	2·11	1·02	50·35	52·24
SA7	1·88	0·70	3·08	43·16	4·30	16·89	15·54	17·88	2·35	8·54	2·50	1·30	36·06	47·77
SA8	1·64	0·84	1·67	46·62	4·88	13·15	19·56	22·55	2·88	6·26	1·01	0·98	44·22	37·32
SA9	1·77	0·83	1·81	47·58	6·00	12·71	17·76	20·26	2·86	6·97	1·08	0·73	39·80	38·06
SA10	1·57	0·93	1·68	48·77	6·68	10·31	18·04	21·10	2·33	7·25	1·01	0·87	46·60	30·60
SA11	2·20	1·05	0·92	50·54	5·04	14·55	16·78	20·27	1·46	5·90	0·40	0·92	25·17	21·64
SA12	2·01	1·03	1·51	57·94	4·00	13·35	7·08	7·56	1·72	12·42	0·81	0·69	31·63	23·14
SA13	2·33	1·16	1·98	59·60	3·70	17·12	5·60	5·53	1·86	9·04	0·55	0·62	20·75	13·44
SA14	2·27	1·04	2·21	59·66	3·74	18·30	4·57	4·16	1·90	9·08	0·48	0·47	22·80	22·39
NS1	1·03	1·27	0·40	35·40	13·00	0·00	23·80	29·65	1·03	18·11	3·05	0·30	63·22	8·63

Table III

SODA BY ANALYSIS AND BY DIFFERENCE

Melt No.	Analysis Na_2O	Difference Na_2O
MA1	39·08	39·79
MA2	33·12	35·23
MA3	20·55	21·85
MA4	6·59	6·11
SA7	18·72	16·89
SA11	15·50	14·55

the total soda content of a slag found by direct determination was not far removed from that calculated by difference :

$$\text{Percentage of soda by difference} = 100 - (FeO + Fe_2O_3 + P_2O_5 + SiO_2 + CaO + MnO + MgO + Al_2O_3)$$

In the results of the chemical analysis of the slags, Tables I and II, the percentage of soda is that found by the difference method.

Little error was introduced by taking the soda as difference, because all the constituents in the slag were accurately determined, and there was little risk of contamination of the slag by other compounds, such as oxides of titanium, chromium, and sulphur. Had these impurities been present, it was considered that they would not have amounted to more than 0·5–1·0%.

To ensure the validity of the percentage of soda by difference, six slags from various series were analysed for soda by Berzelius' method, and, as may be seen from Table III, the percentage of soda by analysis and by difference varied between the maximum limits of ± 2%. Whatever the reason might be for this error, it was within the acceptable limits, and for this particular type of work its accuracy was quite sufficient. One could also deduce from the above fact that the sums of the constituents determined, assuming that the soda was determined for every slag, would fall between 98–102%.

REMARKS ON THE EXPERIMENTAL RESULTS

Series C

Ten melts were made in this series, and magnesia hearths were used. The purpose of these melts was to find out the possible effect of lime on tri-basic sodium phosphate, and the behaviour of soda in basic slags. The main constituents of these slags varied between the following limits :

CaO, %	Na_2O, %	SiO_2, %	P_2O_5, %
47–29	15–8	29–20	14–7

Other constituents were kept low in order to make the relationship between the soda and basic slags more clear. The magnesia contamination of the slags increased as the lime content decreased. The chemical analyses of metal and slag samples of this series are given in Table I.

Series MA

The experimental results on series C led the authors to the conclusion that, although tri-basic sodium phosphate was a very stable compound, it was not possible to keep a reserve of soda, as sodium silicates, in lime-bearing basic slags. This being so, it was decided to use siliceous slags containing soda.

Five melts were made on magnesia hearths with acid slags, and the dephosphorization obtained was very promising. Contamination from the hearth was rather high, and it was considered that there was no reason why silica hearths should not be used.

Series SA

Fourteen melts were made in this series, where acid slags were used on silica hearths. The results on series MA and series SA showed how the oxidizing power of the acid slags varied with composition.

Series NS

Finally, one melt was made with an acid slag on a silica hearth, but without the addition of soda to the slag. This, together with a few other melts in series MA, and series SA, showed the importance of soda in dephosphorization.

The results of the chemical analyses of the last three series are given in Table II. In the synthetic slags of the series MA, SA, and NS, phosphates were not introduced ; the phosphoric acid content of the final slags was the result of dephosphorization.

BEHAVIOUR OF SODA IN BASIC AND ACID SLAGS

Before making an attempt to use soda as a constituent in a suitable slag for the purpose of dephosphorizing iron, it was thought that the proof of the stability of sodium phosphates, in comparison to that of calcium, was essential. It was also desired to find out whether it was possible to stabilize soda, other than in the form of phosphates, in basic slags.

STABILITY OF TRI-BASIC SODIUM PHOSPHATE

In series C the original synthetic slags contained certain amounts of tri-basic sodium phosphate and sodium carbonate, along with lime, silica, and ferric oxide.

The first observation during the slag-making period was the evolution of intense soda fumes. In spite of a heavy loss in soda during the course of the melt, the analysis of the final slags shows the presence of soda varying from 8 to 15%. It is also interesting to note that the amount of soda present in these slags corresponds very closely to that which had been added to the original synthetic slag in the form of tri-basic sodium phosphate. In other words, lime did not appear to affect the tri-basic sodium phosphate, and this could only be due to the fact that tri-basic sodium phosphate was more stable than similar compounds of calcium. If it were not, there would

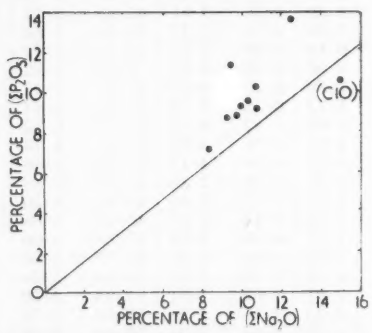


Fig. 4—Relationship between (ΣNa_2O) and (ΣP_2O_5) contents in soda-lime slags

have been replacement of soda by lime, and consequently the freed soda would have been volatilized readily.

It may be noticed, in Table I, that the degree of dephosphorization varies from 8 to 70%, and therefore, the final slags contain more phosphoric acid than was added originally to the slag in the form of tri-basic sodium phosphate. This is shown very clearly in Fig. 4, where the percentage of (ΣP_2O_5) is plotted against the percentage of (ΣNa_2O). The straight line is a theoretical one indicating the amount of phosphoric acid that should be present in combination with soda as a tri-basic phosphate, *i.e.*, $(\Sigma Na_2O)\% = 1{\cdot}31(\Sigma P_2O_5)\%$.

The fact that all the experimental results, with one exception, *i.e.*, C10, lie above this theoretical line indicates that there is an excess of phosphoric acid in the slags which must be combined with lime. An attempt was made to correlate the degree of dephosphorization to the basicity of the slag.

In order to obtain a figure for the effective distribution factor of phosphorus between slag and metal, it was necessary to calculate the amount of phosphoric acid combined with lime. This was done in the following manner :

$$(\Sigma P_2O_5) = (P_2O_5)_{CaO} + (P_2O_5)_{Na_2O} \quad \ldots\ldots\ldots(1)$$

$$\textit{i.e.,}\ (\Sigma P_2O_5) = (P_2O_5)_{CaO} + \frac{(\Sigma Na_2O)}{1{\cdot}31} \quad \ldots\ldots\ldots(2)$$

$$(P_2O_5)_{CaO} = (\Sigma P_2O_5) - \frac{(\Sigma Na_2O)}{1{\cdot}31} \quad \ldots\ldots\ldots(3)$$

where $(P_2O_5)_{CaO}$ is equal to the percentage of phosphoric acid that is combined as tri-calcium phosphate as a result of dephosphorization of the iron. Therefore, the effective distribution factor of phosphorus between the slag and the metal phases may be represented thus :

$$\text{Distribution factor} = \frac{(P_2O_5)_{CaO}}{[\Sigma P]} \quad \ldots\ldots\ldots(4)$$

Finally the basicity ratio, $(B)_1$, may be written as :

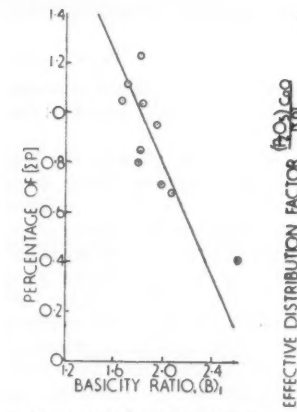
$$(B)_1 = \frac{CaO + FeO + MnO + MgO + Al_2O_3 - 1{\cdot}18\,(P_2O_5)_{CaO}}{SiO_2} \quad \ldots(5)$$

The amounts of soda and phosphoric acid combined as tri-basic sodium phosphate have been neglected, because the phosphate is considered to be stable, and there is no excess soda except in one case, *i.e.*, C10. The results of the calculations are given in Table IV.

Table IV

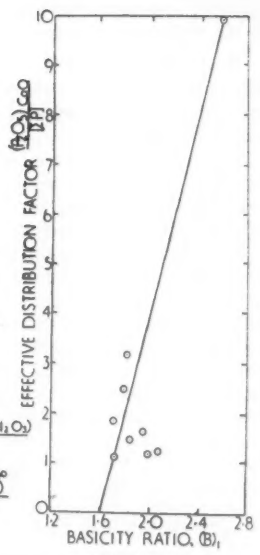
CALCULATED NUMERICAL VALUES ON SODA-LIME SLAGS

Melt No.	P	$(P_2O_5)_{Na_2O}$	$(P_2O_5)_{CaO}$	$\frac{(P_2O_5)_{CaO}}{[\Sigma P]}$	Basicity Ratio $(B)_1$
C1	0·68	6·40	0·71	1·05	2·06
C2	1·04	7·10	1·55	1·49	1·83
C3	0·95	7·70	1·55	1·64	1·94
C4	1·12	7·50	1·26	1·13	1·71
C5	0·71	8·30	0·85	1·20	1·97
C6	0·85	7·94	1·56	1·84	1·70
C7	1·23	9·63	3·87	3·16	1·81
C8	0·80	8·25	1·97	2·47	1·79
C9	0·41	7·25	4·05	9·90	2·60
C10	1·07	10·56	...	...	1·65


PERCENTAGE OF [ΣP]
BASICITY RATIO, (B)₁
(B)₁ = [CaO − 1·18(P₂O₅)CaO] + (FeO) + (MnO) + (MgO) + (Al₂O₃) / (Σ SiO₂)
where, (P₂O₅)Na₂O = (ΣP₂O₅) − (P₂O₅)Na₂O
i.e., (P₂O₅)CaO = (ΣP₂O₅) − (ΣNa₂O) / 1·31

(Fig. 5)

Fig. 5—Relationship between [P]% and $(B)_1$



(Fig. 6)

Fig. 6—Relationship between $\frac{P_2O_5{}_{CaO}}{[\Sigma P]}$ and $(B)_1$

In Fig. 5, the percentage of phosphorus in the iron is plotted against the basicity ratio $(B)_1$, and in Fig. 6, the distribution factor is plotted against the basicity ratio $(B)_1$. Although the number of experimental points is small and they are somewhat scattered, a straight line may be drawn through them in both figures. The indications are that the reduction in phosphorus increases with the basicity ratio $(B)_1$.

Whiteley[4] has shown that there was no dephosphorization when the basicity index fell below two, but it was pointed out that it was most likely that dephosphorization of iron started when the basicity index had a value somewhere between one and two. From the experimental data, as shown in Figs. 5 and 6, it is possible to suggest that dephosphorization of iron starts with the basicity ratio $(B)_1$ between 1·5 and 1·6, which is in good agreement with Whiteley's results.

These curves may be taken as further evidence of the stability of tri-basic sodium phosphate, since the sodium phosphate does not appear to affect the basicity relationships.

BEHAVIOUR OF SODA IN BASIC SLAGS

The most probable compounds of soda in a slag, under steelmaking conditions, are sodium phosphates and sodium silicates. The stability of the former has been established.

When additions of sodium phosphate and sodium carbonate were made to lime-bearing synthetic slags, it was found, at the end of the melt, that the soda remaining in the slag could be accounted for as sodium phosphate. It would appear that all the soda not so combined was removed in the form of a dense white fume. The addition of sodium carbonate to these slags was presumed to cause the formation of sodium silicates, but in the presence of sufficient lime, these

silicates appear to be very unstable, breaking down in the following manner :

$$Na_2O \cdot nSiO_2 + nCaO \rightarrow n(CaO \cdot SiO_2) + Na_2O \quad \ldots\ldots(6)$$

The soda set free by such a reaction volatilized under the temperature conditions prevailing, causing the white fumes.

The unstable nature of sodium silicates in the presence of lime made it obvious that basic slags could not be used as a medium for retaining a reserve of soda for the purpose of dephosphorization.

In order to overcome this difficulty, another series of melts was made under acid slags containing very little lime. The first three melts, *i.e.*, MA1, MA2, and MA3, were made on magnesia hearths. In contrast to series C these slags contained large proportions of soda in the final slag, and in addition, very little fume was observed during the melt. Therefore, so far as stability of sodium silicates is concerned, it would be desirable to use a slag high in silica and low in lime. For such slags it becomes necessary to use silica hearths.

PROCEDURE FOR ACID SLAG-MAKING

Two synthetic slags were prepared as follows :

(1) *First Slag*—This slag corresponded to an early slag in acid steelmaking processes, and its composition varied between the following limits :

SiO_2, %	FeO, %	MnO, %	CaO, %
40–60	20–30	15–25	3–5

(2) *Soda-Oxidizer*—In order to oxidize the phosphorus, and absorb any phosphoric acid formed, a slag containing an oxidant and sufficient soda must be formed. The soda-oxidizer had the following composition range :

Na_2O, %	Fe_2O_3, %	SiO_2, %
30–60	0–50	0–50

The silica was added to adjust the final silica content of the slag to that which was required.

A great advantage was gained by making up the slag in stages. Having made up the first slag, which was highly siliceous, the soda-oxidizer was taken up by the slag very easily with very little loss in soda. During the soda-oxidizer additions a continuous boil took place, and this period may be regarded as the refining period. After the final addition of the soda-oxidizer, steady conditions in the bath were maintained for 10 min. The method of sampling was not changed.

BEHAVIOUR OF SODA IN ACID SLAGS

The results of 20 melts using acid slags are tabulated in Table II.

In Fig. 7, the percentages of (ΣNa_2O) and $(Na_2O)_{SiO_2}$ are plotted against the percentage of (ΣCaO),

where (ΣNa_2O) = percentage of total soda,
$(Na_2O)_{SiO_2}$ = percentage of soda that forms silicates

i.e., $(Na_2O)_{SiO_2} = (\Sigma Na_2O) - 1{\cdot}31(\Sigma P_2O_5)$

Calculated values of $(Na_2O)_{SiO_2}$ are tabulated in Table V.

These two curves, *A* and *B*, show that soda is replaced when the percentage of total lime increases. It is suggested that the scatter of the results is due to the fact that the possible effect of the total silica content of the slag has not been taken into account.

This replacement phenomenon is shown more clearly in Fig. 8, where the percentage of $(Na_2O)_{SiO_2}$ is plotted against the ratio $(\Sigma CaO)/(\Sigma SiO_2)$, (*see* Table V). The experimental points lie more closely on the curve.

When the $(\Sigma CaO)/(\Sigma SiO_2)$ ratio is below 0·07, the amount of $(Na_2O)_{SiO_2}$ increases from 12% up to 35%. As the ratio increases from 0·07 to 0·19 there is a gradual drop in the percentage of $(Na_2O)_{SiO_2}$, and within that limit the curve is a straight line. As the lime/silica ratio increases above 0·19, the percentage of $(Na_2O)_{SiO_2}$ falls from 8% at an increased rate. Finally, at a ratio of 0·26, the amount of $(Na_2O)_{SiO_2}$ falls to zero, and for any ratio above this value no sodium silicate forms. Any soda remaining in the

Table V

CALCULATED NUMERICAL VALUES ON SODA-ACID SLAGS

Melt No.	Wt. of P_2O_5, g.	$(Na_2O)_{SiO_2}$, %	$\frac{(\Sigma CaO)}{(\Sigma SiO_2)}$	Basicity Ratio, $(B)_2$	Reduction, C %	Reduction, P %
MA1	...	35·37	0·04	1·09	20·75	29·11
MA2	...	30·50	0·05	1·11	22·11	28·36
MA3	20·20	8·27	0·19	1·52	51·02	84·34
MA4	19·37	0·00	0·59	2·59	94·27	97·13
MA5	14·15	0·00	1·01	...	98·20	89·20
SA1	17·30	3·68	0·21	1·45	48·99	87·33
SA2	15·00	7·52	0·19	1·44	41·51	81·34
SA3	12·00	8·44	0·17	1·30	33·67	61·20
SA4	10·20	9·31	0·13	1·33	33·67	58·22
SA5	9·50	2·47	0·22	1·10	56·78	43·88
SA6	6·51	0·00	0·26	1·23	50·35	52·24
SA7	7·30	12·85	0·10	1·10	36·06	47·77
SA8	5·35	10·96	0·10	1·00	44·22	37·32
SA9	5·67	10·34	0·12	0·94	39·80	38·06
SA10	4·64	8·69	0·14	0·93	46·60	30·60
SA11	3·12	13·34	0·10	0·91	25·17	21·64
SA12	2·43	11·37	0·07	0·64	31·63	23·14
SA13	4·40	14·53	0·06	0·57	20·75	13·44
SA14	4·87	15·40	0·06	0·56	22·80	22·39
NS1	0·83	0·00	0·37	1·81	63·22	8·63

slag is in the form of tri-basic sodium phosphate, which is not affected by the lime/silica ratio.

The relationships, shown in Fig. 8, are the result of the probable replacement of soda by lime. It is believed that immediately after the addition of the soda-oxidizer, some dephosphorization is brought about, according to the amount of iron oxide available for the oxidation of phosphorus. The remainder of the soda will be in the form of silicates, which apparently decompose readily in the presence of lime, forming more stable silicates. The freed soda, in the absence of any phosphoric acid, will then volatilize. This reaction may be represented by the following equation :

$$Na_2O.nSiO_2 + n(CaO \rightarrow nCaO.SiO_2) + Na_2O \quad \ldots\ldots(7)$$

This reaction is possibly an irreversible one, because free soda is easily volatilized. This process of replacement of soda by lime, therefore, will continue until there is no more sodium silicate, or lime. If the ratio of lime to silica, in the final slag, is below 0·26, some soda will exist in the form of sodium silicate.

The importance of this replacement phenomenon lies in the fact that it answers the question " what kind of slag can hold soda in reserve for dephosphorization ? " Whereas lime in a slag has no effect on sodium phosphates, its presence is detrimental to the reserve of soda essential for dephosphorization. The conditions in high lime-bearing basic slags, therefore, are most unfavourable, if not impossible, for the formation of the sodium silicates necessary as a source of soda for dephosphorization. In basic slags the lime/silica ratio varies from 3 to 4, which is extremely high in comparison with the results shown in Fig. 8. Consequently, it becomes very obvious that soda can be applicable only in acid slags in which the lime/silica ratio does not exceed 0·26, and preferably its value should lie below 0·19.

The above argument applies to the availability of soda only, and takes no cognizance of the oxidizing power of the slag, which is a second criterion for dephosphorization. It so happens that other factors are involved in the low lime/silica ratio slags, which affect their oxidizing power very materially.

OXIDIZING POWER OF SODA-ACID SLAGS

Various investigators have represented the dissociation of fayalite by the following equation :

$$2FeO + SiO_2 \rightleftharpoons 2FeO.SiO_2 \quad \ldots\ldots(8)$$

Under equilibrium conditions its chemical constant may be written as follows :

$$\frac{(FeO)^2 \times (SiO_2)}{(2FeO.SiO_2)} = D_{2FeO.SiO_2} \quad \ldots\ldots(9)$$

where the reactants (FeO) and (SiO_2) are in their free states. The equilibrium constant depends on temperature, and according to H. Schenck and Brüggemann[5] this relationship is given by the following equation :

$$\log D_{2FeO.SiO_2} = -\frac{10500}{T} + 7{\cdot}55 \quad \ldots\ldots(10)$$

Before any use can be made of the last two equations, *i.e.*, (9) and (10), two assumptions are necessary :

(i) It is assumed that at the time of sampling, the fayalite reaction, equation (8), reaches a state of equilibrium

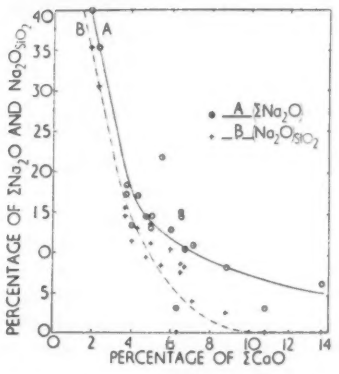


Fig. 7—Effect of lime on soda content of the slag

(ii) It is necessary to consider the other silicate-forming bases, *e.g.*, Na_2O, CaO, MnO, MgO, and Al_2O_3. In acid slags these bases will form silicates rather than liquid solutions between themselves, mainly because of the high silica contents of the slags.

On this basis the following high-silica silicates are assumed to form :

$Na_2O.SiO_2$; $Na_2O.2SiO_2$; $CaO.SiO$; $MnO.SiO_2$; $MgO.SiO_2$; $3Al_2O_3.2SiO_2$.

The compound, $Na_2O.2SiO_2$, is assumed to form in slags containing more than 40–43% SiO_2. By analysis the total percentages of these bases are known, and it then becomes possible to calculate the amount of silica that is required by each base to form a silicate. These calculations are made stoichiometrically, and the stability of the silicate has been assumed for the following reasons :

(i) The amounts of the bases are low, and in addition, the percentage of silica is high

(ii) The temperature is low, *i.e.*, 1400° C., when the dissociation of the silicates may be regarded as practically negligible.

With these assumptions it is possible to write an equation for silica :

$$|SiO_2|_{Base} = (SiO_2)_{Na_2O} + (SiO_2)_{CaO} + (SiO_2)_{MnO} + (SiO_2)_{MgO} + (SiO_2)_{Al_2O_3} \quad \ldots\ldots(11)$$

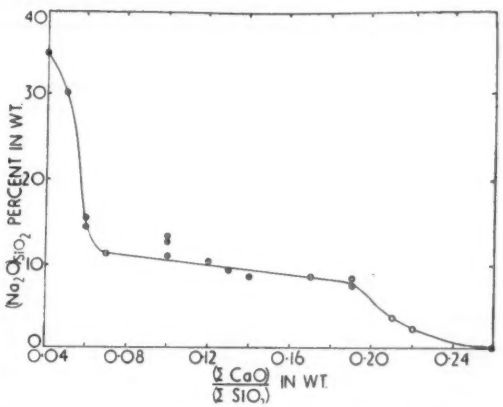


Fig. 8—Effect of the ratio $(\Sigma CaO)/(\Sigma SiO_2)$ on percentage of $(Na_2O)_{SiO_2} = (\Sigma Na_2O) - (Na_2O)_{P_2O_5}$

$$i.e., |SiO_2|_{Base} = \frac{(Na_2O)_{SiO_2}}{1{\cdot}03} + \frac{(CaO)}{0{\cdot}93} + \frac{MnO}{1{\cdot}18} + \frac{(MgO)}{0{\cdot}67} + \frac{(Al_2O_3)}{1{\cdot}70} \quad \text{.....(12)}$$

$$\text{where } (Na_2O)_{SiO_2} = (\Sigma Na_2O) - (Na_2O)_{P_2O_5} = (\Sigma Na_2O) - 1{\cdot}31 \times (\Sigma P_2O_5) \quad \text{...(13)}$$

and when the compound, $Na_2O.2SiO_2$, forms :

$$(SiO_2)_{Na_2O} = \frac{(Na_2O)_{SiO_2}}{0{\cdot}516}$$

By considering the silica that takes part in the fayalite reaction, the silica equation is completed thus :

$$(\Sigma SiO_2) = (SiO_2) + (SiO_2)_{FeO} + |SiO_2|_{Base} \quad \text{...(14)}$$

where (ΣSiO_2) = percentage of total silica as analysed

(SiO_2) = percentage of free silica that takes part in the fayalite reaction,

$(SiO_2)_{FeO}$ = percentage of silica that is in the form of fayalite,

$|SiO_2|_{Base}$ = percentage of silica that combines with other bases, see equations (11) and (12).

Since the amounts of (ΣSiO_2) and $|SiO_2|_{Base}$ are known by analysis, and by calculation, respectively, equation (14) can be simplified in order to calculate the sum of the free silica and the silica that is in the form of fayalite.

$$i.e., \{SiO_2\} = (SiO_2) + (SiO_2)_{FeO} = (\Sigma SiO_2) - |SiO_2|_{Base} \quad \text{...(15)}$$

Since the amounts of (ΣFeO), $\{SiO_2\}$ and $D_{2FeO.SiO_2}$ are known, for a given temperature, it is possible to write a mathematical equation to calculate the amounts of the free constituents of the fayalite reaction that are at equilibrium :

Let $(FeO) = X$

and $(2FeO.SiO_2) = (\Sigma FeO - X) \times 1{\cdot}418$

and $(SiO_2)_{FeO} = \frac{(\Sigma FeO - X)}{2{\cdot}39}$

where (FeO) is in a free state and (ΣFeO) is the total ferrous iron as analysed. From equation (15) :

$$(SiO_2) = \{SiO_2\} - (SiO_2)_{FeO} = \{SiO_2\} - \frac{(\Sigma FeO - X)}{2{\cdot}39}$$

These equivalents may be substituted in equation (9), *i.e.*, the equation of the fayalite dissociation constant :

$$\frac{(FeO)^2 \times (SiO_2)}{(2FeO.SiO_2)} = D_{2FeO.SiO_2} = \frac{X^2\left[\{SiO_2\} - \frac{(\Sigma FeO - X)}{2{\cdot}39}\right]}{1{\cdot}418 \times (\Sigma FeO - X)} \quad \text{.....(16)}$$

The equation may be simplified as follows :

$$X^3 + \left[\{SiO_2\}\,2{\cdot}39 - (\Sigma FeO)\right]X^2 + (1{\cdot}418 \times 2{\cdot}39 \times D_{2FeO.SiO_2})X - (1{\cdot}418 \times 2{\cdot}39 \times D_{2FeO.SiO_2}) \times (\Sigma FeO) = 0 \quad \text{...(17)}$$

The variation of the equilibrium constant, $D_{2FeO.SiO_2}$, as mentioned before, has been represented by the following equation :

$$\log D_{2FeO.SiO_2} = -\frac{10500}{T} + 7{\cdot}55$$

Since the experiments were carried out at 1400° C. ± 10 this becomes

$$\log D_{2FeO.SiO_2} = -\frac{10500}{273 + 1400} + 7{\cdot}55 = 1{\cdot}2740$$

$$D_{2FeO.SiO_2} = 18{\cdot}80 \quad \text{.....(18)}$$

Putting this value in equation (17), and by further simplification it follows that :

$$X^3 + \left[\{SiO_2\}\,2{\cdot}39 - (\Sigma FeO)\right]X^2 + 63{\cdot}7X - 63{\cdot}7(\Sigma FeO) = 0 \quad \text{.....(19)}$$

By means of equation (19), the percentage of free ferrous oxide may be calculated, from which also the other free constituents of the fayalite reaction in equilibrium can be found. Calculations on the experimental data are given in Table VI.

In general, the dissociation constant, $D_{2FeO.SiO_2}$, increases with temperature. As the total silica decreases, or as the basicity increases, the extent of

Table VI

RESULTS OF CALCULATIONS ON FAYALITE REACTION

Melt No.	(ΣFeO), %	$(FeO)_{SiO_2}$, %	$(FeO)_{free}$, %	(ΣSiO_2), %	$\{SiO_2\}$, %	$(SiO_2)_{free}$, %	$2FeO.SiO_2$, %	$\frac{(2FeO.SiO_2)}{(SiO_2)}$
MA1	2·88	0·35	2·53	43·20	1·54	1·39	0·50	0·36
MA2	6·17	2·20	3·97	42·56	4·64	3·72	3·12	0·84
MA3	6·95	0·85	6·10	28·24	0·96	0·60	1·21	2·01
SA1	26·10	16·33	9·77	33·96	11·68	4·83	23·16	4·79
SA2	22·22	12·07	10·15	34·30	8·18	3·13	17·11	5·47
SA3	22·12	14·32	7·80	37·60	12·26	6·27	20·03	3·24
SA4	24·50	16·41	8·09	37·68	13·56	7·03	23·26	3·31
SA5	21·66	14·96	6·70	40·42	15·13	8·88	21·16	2·38
SA6	29·25	21·73	7·52	40·70	19·33	10·23	30·81	3·01
SA7	17·88	12·00	5·88	43·16	14·28	9·24	17·02	1·84
SA8	22·55	14·82	7·73	46·62	12·81	6·61	21·01	3·18
SA9	20·26	13·26	7·00	47·58	12·79	7·24	18·80	2·60
SA10	21·10	14·49	6·61	48·77	14·90	8·83	20·54	2·33
SA11	20·27	13·43	6·84	50·54	13·25	7·62	19·04	2·50
SA12	7·56	4·83	2·73	57·94	19·48	17·46	6·85	0·32
SA13	5·53	3·24	2·25	59·60	18·58	17·21	4·65	0·27
SA14	4·16	2·24	1·92	59·66	17·04	16·10	3·17	0·20
NS1	29·65	16·06	13·59	35·40	9·04	2·32	22·76	9·81

the dissociation of fayalite increases at a constant temperature. Also, the greater the amount of total iron present the greater the amount of fayalite, and therefore, the greater the amount of free ferrous iron as a result of dissociation.

It can, therefore, be deduced that if the slag is at a high temperature during the metal refining period, and contains low silica and high total iron, the amount of ferrous oxide available for the oxidation of carbon and phosphorus in iron will be an optimum value. In other words, under these three conditions an acid slag will attain a high oxidizing power.

OXIDATION OF PHOSPHORUS

In all steelmaking processes, except the Bessemer, the oxygen is supplied to the metal through the slag. The chemical and physical conditions of the slag are controlling factors in determining the distribution of the oxygen between the slag and metal. Oxygen is supplied to the slag either by gas oxidation or direct iron oxide additions, and in most cases by both.

At the slag–metal interface ferric oxide is reduced to ferrous oxide by iron. The ferrous oxide can then function in two ways ; firstly, it can combine with silica to form fayalite, and secondly, it can pass into solution in the slag. The oxide that forms fayalite is considered not to take part in the distribution of oxygen between the slag and metal, whereas it is the oxide in solution, the so-called ' free ' ferrous oxide, that supplies oxygen to the metal. Therefore, for the same amount of oxide addition to the slag during the melt, most oxygen will be supplied to the metal when fayalite formation is least. This, however, may be formulated as follows :

Let $\{Fe\}$ be the total amount of ferrous iron supplied to the slag during a melt, from start to finish,

$(Fe)_c$ be the amount of ferrous oxide that had been required for decarburization,

$(Fe)_p$ be the amount of ferrous oxide that had been required for dephosphorization,

(ΣFe) be the amount of total iron in the final slag as FeO,

then,

$$\{Fe\} = (Fe)_c + (Fe)_p + (\Sigma Fe) \quad \ldots\ldots\ldots(20)$$

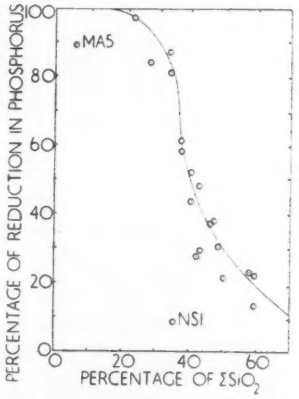


Fig. 9—Effect of total silica content of the slag on dephosphorization

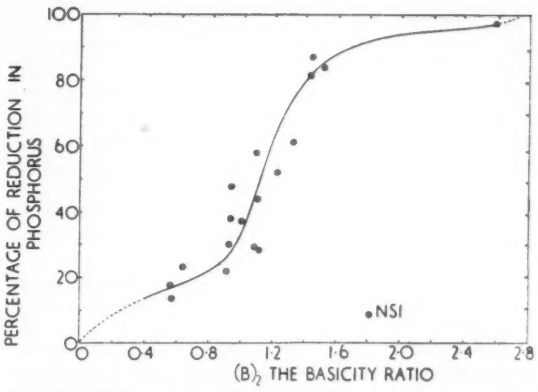


Fig. 10—Effect of basicity ratio of soda-acid slags on dephosphorization

Incidentally, the presence of other accompanying elements in iron, *e.g.*, manganese and silicon, have been ignored, because they were present in small percentages only. The amount of iron oxide used up for decarburization and dephosphorization, therefore, will be :

$$(Fe)_c + (Fe)_p = \{Fe\} - (\Sigma Fe) \quad \ldots\ldots\ldots(21)$$

It would have been very interesting if the experimental data could have shown the validity of equation (21), but unfortunately, the exact value of $\{Fe\}$, which is equal to the iron oxide addition to the synthetic slag plus gas oxidation, was not known. Although the experimental melts were made in an atmosphere of nitrogen, gas oxidation was inevitable.

Assuming the validity of equation (21), it follows that for the same value of $\{Fe\}$ there will be more decarburization and dephosphorization if the final total iron content of the slag is low. Since the products of oxidation of the carbon and phosphorus are gaseous and stable, respectively, the extent of the oxidation is determined by the equilibrium conditions for the fayalite dissociation.

In Fig. 9, the percentage reduction of phosphorus is plotted against the percentage of total silica in the slag. It shows that the degree of dephosphorization increases as the total silica content of the slag decreases. As the total silica content of the slag decreases from 40% to 30%, the reduction in phosphorus increases from 50% to 90%. It should be mentioned that an excess of ferric oxide was added in each melt.

In Fig. 10, the percentage reduction of phosphorus is plotted against the basicity ratio, $(B)_2$, of the slag (*see* Table V) where :

$$(B)_2 = \frac{(Na_2O)_{SiO_2} + FeO + CaO + MnO + MgO + Al_2O_3}{SiO_2} \quad \ldots(22)$$

It shows an increase in dephosphorization with basicity. In arriving at the basicity ratio, $(B)_2$, the P_2O_5 and the Na_2O combined with it have not been taken into account, since it is considered that the tri-basic sodium phosphate is stable under the condition of experiment.

The curves in Figs. 9 and 10 may suggest that it is necessary to use slags of high basicity ratio for dephosphorization.

The basicity ratio as defined previously is rather different from the normally accepted use of the term in basic steelmaking practice, since the bases other than soda and ferrous oxide are low. Its value is determined to a great extent by the total silica content of the slag. The introduction of lime into the slag for the purpose of increasing the basicity ratio is undesirable, since it removes soda, and one of the conditions of dephosphorizing, as has already been mentioned, is an excess of soda retained in the slag as a silicate.

The effect of silica in the slags is twofold :

(i) An increase in silica reduces the basicity ratio
(ii) An increase in silica stabilizes fayalite at a given temperature.

The net effect of an increase in silica at the temperatures employed in this research is to cause less oxygen to be supplied to the metal for the purpose of dephosphorizing. Thus, at low values of the basicity ratio the degree of dephosphorization should be less. Provided sufficient soda is retained in the slag, the higher the basicity ratio the better, since, under these conditions, more fayalite will dissociate.

In Fig. 11, the ratio $\frac{2FeO\,.\,SiO_2}{SiO_2} = \frac{(FeO)^2}{D_{2FeO\,.\,SiO_2}}$ is plotted against the percentage reduction of phosphorus, where (FeO) and (SiO_2) are the concentrations of free ferrous oxide and free silica in the slag, ($2FeO\,.\,SiO_2$) is the fayalite content of the slag, and $D_{2FeO\,.\,SiO_2}$ is the dissociation constant of the fayalite reaction, which is 18·80 at 1400° C. The method of calculation of these constituents has already been discussed, and the results of the calculations are given in Table VI. Figure 11 shows four curves drawn through the experimental points for different ranges of total iron content of the slag, and also, six parallel lines drawn to indicate the regions of equal silica content of the slags within which the experimental points fall. Considering the possible experimental errors, and also the assumption made in the calculation of the constituents of the fayalite reaction, the scatter of the points is not too great.

In order to show the significance of the curves in Fig. 11, the same graph is redrawn in its schematic form in Fig. 12. The characteristics of the curves may be sub-divided as follows :

(1) *Decreasing* (ΣSiO_2), *and* {Fe} *Remaining Constant*—The decrease in silica from *A* to *D* and/or a rise in basicity causes a greater dissociation of fayalite, *i.e.*, less ferrous oxide is stabilized as fayalite, and, as a result, more ferrous oxide becomes available for the purposes of oxidation. Furthermore, since there is no change in the amount of {Fe}—*see* equation (20)—more dephosphorization and decarburization causes the total iron to fall from *A* to *B*. The resultant *AC* shows the direction in which the point *A* travels when the total silica content of the slag falls.

(2) *Decreasing* (ΣSiO_2), %, *and Increasing* {Fe} —By increasing the amount of iron oxide from *A* to *B*, at the same time reducing the total silica content of the slag from *A* to *D*, the point *A* moves to point *C'*, and the final total iron content of the slag is about the same as at *A*. Furthermore, on comparing sections 1 and 2 it is seen that the ratio ($2FeO\,.\,SiO_2$)/(SiO_2) and the percentage reduction of phosphorus are higher for the points in the direction of *AC'* than for those in the direction of *AC*, for the same decrease in the total silica content of the slag.

(3) (ΣSiO_2) *Remaining Constant, and Increasing* {Fe}—There are two factors involved. Firstly, an increase in the amount of iron oxide added must mean a greater quantity of fayalite formed, which in turn, given constant dissociation conditions, must produce a greater quantity of free ferrous oxide, which may bring about greater degrees of dephosphorization. Secondly, according to the equation for the fayalite reaction in equilibrium, the greater the ferrous oxide, and the greater the quantity of fayalite formed, the lower

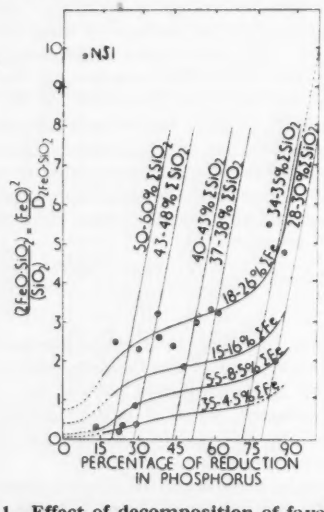


Fig. 11—Effect of decomposition of fayalite on dephosphorization

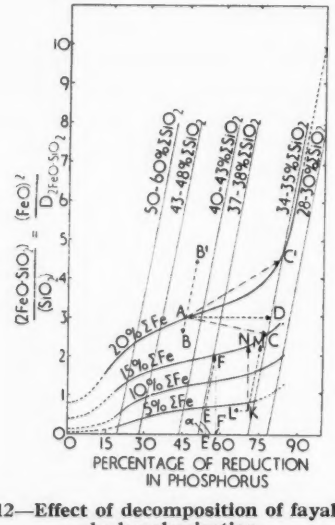


Fig. 12—Effect of decomposition of fayalite on dephosphorization

will be the free silica content of the slag. Consequently, under the conditions of section 3 the percentage of phosphorus reduction increases from *E'* to *F'* and, at the same time, the ratio $(2FeO \cdot SiO_2)/(SiO_2)$ increases from *E* to *F*.

(4) *Increasing* (ΣSiO_2), *and Increasing* {Fe}—If silica increases from *K* to *L*, and total iron increases from *K* to *M*, greater quantities of fayalite are formed. Under certain circumstances, the dissociation of this greater quantity of fayalite may produce greater quantities of ferrous oxide for oxidation of the melt. The quantity of ferrous oxide formed depends on the relative increase of total silica and {Fe}. The resultant of these two changes may be shown by *KN*. Under perfectly balanced conditions the resultant *KN* may be vertical to the abscissa, *i.e.*, there is no increase in dephosphorization.

It has been shown that under the conditions of sections 3 and 4, which are concerned with increasing additions of {Fe}, the increase in the percentage of dephosphorization is not very high at the temperature employed. This is assumed to be due to the fact that a large proportion of the increased addition of iron oxide is consumed in the formation of fayalite. It seems that at a temperature of 1400° C., the most effective method of obtaining more iron oxide for oxidation purposes is to decrease the total silica content of the slag.

The Effect of Temperature

In the present work only two melts, *i.e.*, MA4 and MA5, were finished at 1550° C. These two melts were fortuitous, and unfortunately the slags contained silica contents which were lower than the average, but the dephosphorization was in the order of 90% (*see* Table II). It may be considered that the effect of temperature on the dephosphorization reaction ought not to be measured on these two melts alone.

It is very well known that fayalite decomposes as the temperature increases under constant slag conditions, and the temperature–dissociation constant relationship is given in equation (10). It is presumed that if the temperature of the bath is kept in the order of 1550° C., the extent of the fayalite dissociation will be great, and, as a result, larger quantities of ferrous oxide will become available for the oxidation of the phosphorous to P_2O_5 which, in the presence of soda, will be removed to the slag.

Since the dissociation of fayalite is not only dependent on temperature but must be affected by the silica content of the slag, the slope of the iso-silica lines, in Fig. 12, should be considered. For example, the slope of the line *EF* may be written as :

$$\tan\alpha = \frac{\text{Increase in } \dfrac{(2FeO \cdot SiO_2)}{(SiO_2)}}{\text{Increase in phosphorus reduction, \%}} \quad \ldots\ldots(23)$$

or it may be written in a more general way as :

$$\tan\alpha = \frac{\dfrac{(2FeO \cdot SiO_2)}{(SiO_2)}}{\text{Phosphorus reduction, \%}} \quad \ldots\ldots\ldots\ldots\ldots(24)$$

When the temperature increases, the ratio $(2FeO \cdot SiO_2)/(SiO_2)$ decreases, because of the dissociation of fayalite, and, as a result, more dephosphorization takes place. Reference to equation (24) indicates that, as the temperature increases, the slope of the iso-silica lines (Fig. 12), *i.e.*, tan α, decreases. The angle α of the iso-silica lines is independent of the slag composition, but decreases as the temperature increases, and therefore, for the same temperature these lines are parallel to one another. Furthermore, the reciprocal of the angle α is proportional to the increase in the percentage of phosphorus reduction.

It may be considered, therefore, that the characteristics of Fig. 11, as well as those of Figs. 9 and 10, confirm the theory that dephosphorization increases as the conditions of temperature and slag composition favour fayalite dissociation. There are two important factors which control the decomposition of fayalite and, consequently, the oxidation of phosphorus :

(i) At low temperatures the total silica content of the slag must be low
(ii) With slags of high total silica contents the temperature must be high.

GENERAL DISCUSSION

Generally speaking, the dephosphorization of iron under lime-bearing slags may be considered to take place, firstly, by the oxidation of the phosphorus to phosphoric acid, and secondly, by the stabilization of the phosphoric acid in the slag as a calcium phosphate. It is very well known that the extent of the removal of phosphorus is high if :

(i) The total iron content of the slag is high
(ii) The total lime content of the slag is high
(iii) The total silica content of the slag is low
(iv) The total phosphoric acid content of the slag is low
(v) The temperature is low.

The process of dephosphorization is limited by equilibrium conditions, which are characterized by the compositions of the slag and metal in relation to temperature.

In the original series of melts, series C, which consisted of melts made under slags containing lime and soda, there was found to be insufficient excess soda to bring about dephosphorization. The soda remaining in these slags could be accounted for by that which had been added as tri-basic phosphate. The soda added as carbonate was removed as explained earlier. The dephosphorization brought about in these slags, therefore, must have been due to the excess of lime that was present. Although the number of experiments carried out was small, there was a definite indication that the dephosphorization progressed according to the accepted principles as shown in Figs. 5 and 6. As a result of these experiments, it was confirmed that sodium phosphate was more stable than calcium phosphate.

In the case of dephosphorization of iron by soda-acid slags the situation is different, since the sodium phosphate formed is stable. The dephosphorizing reactions are then independent of the final temperature and the final slag composition. They are only affected by the temperature and composition of the slag during the refining period, which control, firstly, the oxidizing power of the slag, and secondly, the reserve of soda.

It has already been pointed out that the oxidizing power of the slag is determined by the extent of the dissociation of the fayalite, which is controlled by

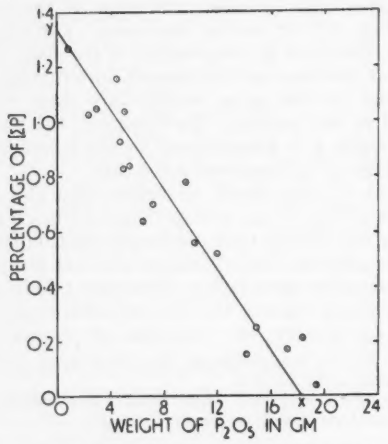


Fig. 13—Relationship between [P]% and weight of P_2O_5 in the slag

temperature and slag composition. At 1400° C., the temperature at which these experiments have been carried out, the fayalite dissociation in silica slags may reach equilibrium conditions, and, as a result, the slag ceases to be oxidizing. It is necessary, therefore, during the refining period, to maintain slags that allow the dissociation to take place. That is, at a temperature of 1400° C. the slags must contain less than 40% silica (*see* Fig. 9).

It is probable that at higher temperatures an equivalent amount of dissociation could be brought about with increased silica percentages in the slags, in which instance the curve shown in Fig. 9 will be displaced to the right.

If the slag, during the refining period, is of the correct composition and temperature, which enables it to be highly oxidizing, and at the same time it contains sufficient soda, a high degree of dephosphorization will be certain.

The oxidation of phosphorus and the formation of tri-basic sodium phosphate may be represented by the following equations :

$$2P + 5FeO \rightarrow P_2O_5 + 5Fe \quad \text{.....(25)}$$

$$3(Na_2O \cdot nSiO_2) + P_2O_5 \rightarrow 3Na_2O \cdot P_2O_5 + 3nSiO_2 \quad (26)$$

Since tri-basic sodium phosphate is a stable compound, both reactions in equations (25) and (26) are practically irreversible, and they proceed until one of the reactants is reduced to a very low value.

It was considered necessary to show that the amount of phosphorus removed from the metal could be accounted for by the amount of phosphoric acid in the slag. In order to assess the weight of the phosphoric acid in the slag, it was essential to know the slag bulk. This was obtained to a first approximation by considering the manganous oxide contents of the initial and final slags. For example :

$$\text{Slag Bulk} = \frac{\text{Weight of MnO put in the initial slag}}{\text{Percentage of MnO in the final slag}} \times 100 \text{ g.} \quad \text{...(27)}$$

$$\text{Weight of } P_2O_5 = \frac{\text{Slag bulk}}{100} \times (\Sigma P_2O_5)\% \text{ g.} \quad \text{...(28)}$$

The weights of phosphoric acid are given in Table V, and the weight of metal used was 600 g. in each case.

In Fig. 13, the percentage of phosphorus in the iron is plotted against the total weight of phosphoric acid in the slag. The experimental points are evenly distributed on both sides of a straight line which is drawn through the theoretical points X and Y so that :

X = 18·41 g. P_2O_5 for 100% dephosphorization
Y = 1·34% P for 0·00% dephosphorization.

The divergence of the points from the straight line is not great, considering the approximate method of calculating the slag bulk. It confirms the fact that the dephosphorization is solely a slag–metal reaction, and that the experimental results are within the limits of experimental error.

A relationship may be shown between the phosphoric acid and silicate-forming soda contents of the slag. In Fig. 14, the percentage of $(Na_2O)_{SiO_2}$ is plotted against the percentage of P_2O_5. Separate curves have been drawn through the experimental points for narrow ranges of total soda contents, and these curves are parallel to one another. Since the phosphate reaction is irreversible, the $(Na_2O)_{SiO_2}$ content of the slag decreases as the percentage of P_2O_5 increases. Therefore, for a given amount of total soda the $(Na_2O)_{SiO_2}$ decreases and the (ΣP_2O_5) increases as the oxidizing power of the slag increases.

As already mentioned, the amount of soda that forms tri-basic sodium phosphate may be found stoichiometrically, *i.e.* :

$$(Na_2O)_{P_2O_5} = 1{\cdot}31(\Sigma P_2O_5) \text{.....(29)}$$

In the majority of melts examined the amount of soda in the slags has been sufficient to combine with all the phosphoric acid formed. Since a reserve of soda as a silicate is fundamental to dephosphorization under the present conditions of experiment, a few melts were made under slags containing insufficient soda to combine with the phosphoric acid formed. Four melts which contained no sodium silicate are shown in Table V ; *i.e.*, these melts conform to the following statements :

$$(Na_2O)_{SiO_2} = 0$$

$$\text{or } (\Sigma Na_2O) = 1{\cdot}31(\Sigma P_2O_5) \text{ approx.}$$

The effect of soda on dephosphorization may be summarized as follows :

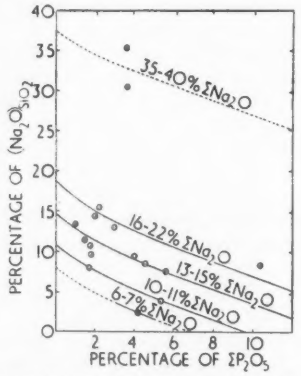


Fig. 14—Relationship between $(Na_2O)_{SiO_2}$ and (ΣP_2O_5) percentages

(1) When the final slag composition shows that the percentage of $(Na_2O)_{SiO_2} > 0$, the extent of dephosphorization during the melt is controlled by the oxidizing power of the slag.

(2) When the final slag composition shows that the percentage of $(Na_2O)_{SiO_2} = 0$, either (*a*) the dephosphorization had been ideal in so far as exactly the correct quantity of soda had been added to combine with all the phosphoric acid that could be formed (this is an improbable and fortuitous condition), or (*b*) the dephosphorization had been partial through lack of soda. Under these circumstances, which are probable, the carbon would be removed in preference to the phosphorus.

In Fig. 15, the percentage reduction of phosphorus is plotted against the percentage reduction of carbon. The curve, drawn through the experimental points, has a peculiar shape for which the authors have no explanation at present. The curve shows that, in the majority of melts, decarburization and dephosphorization take place together. In the absence of a sufficient amount of soda, *i.e.*, $(Na_2O)_{SiO_2} = 0$, the rest of the available iron oxide will be used up by carbon, so that the degree of decarburization will be greater than normal when compared with the degree of dephosphorization. This fact is illustrated by the experimental points MA5, SA5, SA6, and NS1, where the $(Na_2O)_{SiO_2}$ content in the slag is zero, except in SA5 in which $(Na_2O)_{SiO_2}\% = 2{\cdot}47$. This may be due either to undue oxidation of carbon during the melt or an error in analysis.

The melt MA5 was carried out in the absence of a nitrogen atmosphere after the first slag had formed. Incidentally, the melt was made on a magnesia hearth in order to keep the silica content of the slag as low as possible. The gas oxidation was extremely high as may be seen from the final slag composition in Table II. Had there been sufficient soda in the slag, the extent of the removal of phosphorus should have been nearer 99% than the observed 89%, in comparison to the extent of the decarburization. This also explains why the point MA5, in Fig. 9, is well below the curve. Similar conditions apply to melts SA5, SA6 and NS1.

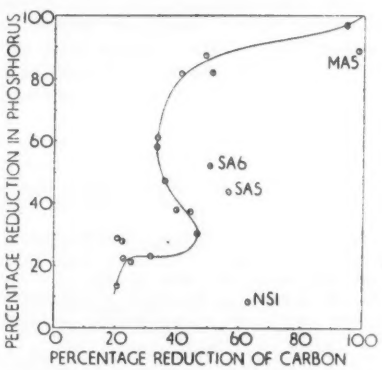


Fig. 15—Relationship between percentage of dephosphorization and percentage of decarburization

A melt, NS1, was made without introducing any soda to the slag. The degree of decarburization indicates (*see* Fig. 15) that the slag was oxidizing in character, but the removal of phosphorus was low, 8·63%. The 8·63% reduction of phosphorus may be attributed to the formation of tri-calcium and/or iron phosphates. Had the slag contained a sufficient amount of soda, the reduction in phosphorus would have been between 80% and 90%, as shown in Figs. 9, 10, 11, and 15.

Therefore, to dephosphorize iron satisfactorily an oxidizing-acid slag, which contains a sufficient amount of soda, must be used. The minimum amount of soda required for this purpose is :

$$\Sigma Na_2O \text{ by weight} = 1{\cdot}31 \times \Sigma P_2O_5 \text{ by weight} \quad \ldots(30)$$

It is noticeable from Table II that some slags, under which the greatest extent of dephosphorization takes place, contain only small percentages of soda (in the order of 3%–6%). This may be explained by the fact that an increase in basicity and/or the lime/silica ratio will bring about greater phosphorus removal, because, under these conditions, fayalite dissociates to a greater extent, but, at the same time, an increase in the lime/silica ratio must cause a decrease in the $(Na_2O)_{SiO_2}$ content of the slag.

To dephosphorize at low temperatures soda-acid slags must lie within a critical range of composition. For instance, to ensure a good reserve of soda in a slag, the total silica content of the slag must be increased, or the lime/silica ratio must be decreased. These conditions reduce the oxidizing power of the slag, and consequently an excess of soda in the slag cannot be used for dephosphorization. On the other hand, the oxidizing power of the slag may be increased by increasing the lime/silica ratio and decreasing the total silica content of the slag. This may cause a heavy loss in soda. Within these limits dephosphorization of iron at low temperatures becomes possible.

For a predetermined degree of dephosphorization a range of slag composition may be recommended, at 1400° C., for

$$50 \rightarrow 90\% \text{ reduction in phosphorus}$$
$$25 \rightarrow 60\% \text{ reduction in carbon,}$$

and a soda-acid slag, whose composition lies between the following limits during the refining period will be required :

40–30% ΣSiO_2
1·2–1·5 basicity ratio $(B)_2$
0·2–0·3 $(\Sigma CaO)/(\Sigma SiO_2)$
3–6% ΣNa_2O

A sufficient amount of iron oxide must be supplied during the melt for the purpose of oxidation. If, on the other hand, it is required to work with soda-acid slags in which the acidity is very high, *i.e.*, 55–60% SiO_2, it becomes necessary to increase the temperature of the bath so that fayalite dissociates.

CONCLUSIONS

(1) An endeavour has been made to introduce soda into lime-bearing basic slags. Little or no success attended the endeavour, since lime replaced soda in the silicates present, with the consequent volatilization of the freed soda.

(2) The addition of tri-basic sodium phosphate to basic slags established the fact that it was stable in

the presence of lime, and was, therefore, more stable than calcium phosphate.

(3) In order to make available a reserve of soda during dephosphorization sodium silicates must be present.

(4) To ensure the presence of sodium silicates, acid slags were used and melting conducted on silica hearths.

(5) At 1400° C., and in the presence of an excess of soda, the dissociation of fayalite providing free FeO in the slags is the prime factor affecting dephosphorization.

(6) Conditions of temperature and slag composition have been suggested for the production of oxidizing slags.

(7) Under favourable conditions, 85% of the phosphorus present in a 1·34% phosphorus pig iron has been removed under slags containing 30–35% silica on silica hearths at 1400° C.

(8) Indications suggest that, at higher temperatures, an equivalent amount of dephosphorization can be obtained with slags containing greater concentrations of silica.

(9) In this part of the research, only the process of dephosphorization under soda-acid slags, has been studied. There are other probable effects of soda additions, particularly on the removal of sulphur, which have yet to be examined.

POSSIBLE PRACTICAL APPLICATIONS

There would appear to be a number of possible practical applications of the use of soda-acid slags. It will be appreciated that in making suggestions the authors are fully aware that there may be considerable differences in technique, and possibly in results obtained, in trying to reproduce in practice the conditions existing in an experimental furnace.

Since the work carried out in the first part of this research has been conducted at, or near, 1400° C., the use of soda-acid slags in molten-iron mixers would seem to be most profitable. It has been shown that considerable quantities of phosphorus can be removed from phosphoric iron under slags containing 30–35% SiO_2 and up to 6% soda. It is realized that, at the same time, at least in the small experimental furnace, carbon is also removed. Under ideal conditions the rate of removal of phosphorus is about twice that of carbon. In large mixers this figure may well be considerably greater. It does not seem to be beyond the realm of possibility, or industrial ingenuity, to be able to remove 50% of the phosphorus from a phosphoric iron in a mixer, and yet have sufficient carbon in the iron to allow the subsequent steelmaking process to be operated efficiently. If this were found to be practicable, there would appear to be a great advantage in the increased rate of production possible in the steelmaking process.

Some modification in the refractory lining of the hearths may be necessary, since the soda-acid slags operate best in the absence of lime and are retained with little corrosion in siliceous hearths.

Although the effect of temperature has not been studied in detail, from the indications obtained from the few melts at 1550° C., which confirm theoretical considerations, soda-acid slags containing over 50% silica should bring about dephosphorization, at steelmaking temperatures, in acid-lined furnaces.

Since discussion with Mr. H. H. Burton and his melting-shop colleagues at the English Steel Corporation, large-scale experiments in acid-lined steelmaking furnaces have been carried out; when sufficient data has accumulated the results will be reported.

One of the difficulties in these experiments has been to produce oxidizing conditions in slags that contain fayalite at 1400° C. It is suggested that if oxygen were supplied by air, as in the Bessemer processes, this difficulty would disappear. The introduction of soda into acid Bessemer slags immediately after the first slags form would appear to be an industrial experiment that would be worth while. The application of synthetic soda-acid slags after the manner of Perrin would also appear to be possible.

FUTURE WORK

The present paper is factual in many respects, and there are many problems associated with the work which it is proposed to do in future :

(1) The petrographic and X-ray examination of the soda-acid slags, and the formulation of their constitution is a necessity.

(2) The effect of temperature, particularly in the range of steelmaking temperatures, will be examined with especial reference to slags containing high silica contents.

(3) It has already been mentioned that this section of the work has been concerned with the removal of phosphorus only. There are many other probable effects, particularly that on the removal of sulphur, which require intensive investigation.

(4) A detailed examination of the effects of soda-acid slags on refractories appears to be a necessity if they are to be used in practice.

Acknowledgments

The authors wish to thank Mr. H. H. Burton and his colleagues for their assistance, and also the management of the English Steel Corporation for the facilities which were placed at their disposal for carrying out experimental melts. The confidence in the work which this co-operation implied was particularly gratifying. The authors wish to express their appreciation for the many helpful discussions which they have had with their colleagues at Sheffield University.

References

1. K. Balajiva, A. G. Quarrell, and P. Vajragupta : *Journal of The Iron and Steel Institute*, 1946, No. I, pp. 117–118.
2. C. R. Taylor and J. Chipman : *Transactions of the American Institute of Mining and Metallurgical Engineers*, 1943, vol. 154, p. 228.
3. J. W. Mellor : " A Treatise on Quantitative Inorganic Analysis," pp. 219–221. London, 1913, Charles Griffin and Co.,
4. J. H. Whiteley : *Proceedings of the Cleveland Institution of Engineers*, Session 1922–23, No. 2, p. 36.
5. H. Schenck and E.-O. Brüggeman : *Archiv für das Eisenhüttenwesen*, 1935–36, vol. 9, Jan. 9, p. 43.

Characteristics of Metallographic Polishing Powders

By E. C. Rollason, M.Sc., Ph.D., F.I.M., E. Sharratt, B.Sc., Ph.D., F.R.I.C., and R. R. Roberts, M.A., L.I.M.

SYNOPSIS

The physical properties of powders used for metallographic polishing of mild-steel specimens have been studied. The most essential factor for a successful polishing medium is that the particles shall be plate-shaped, and it is suggested that the polishing depends largely on the frictional characteristics between the flat plates of the polishing powder and the specimen, and not on a cutting action.

Gamma alumina as obtained by the calcination of alum at 1100° C. for 1 hr. consists of such particles, but alpha alumina may also be obtained in the form of plate-shaped particles suitable for polishing purposes by the calcination of ammonium alum at 1250° C. for 1 hr. Normal commercial samples of alpha alumina are, however, granular, and not suitable for polishing purposes.

Gamma alumina is softer than the alpha variety and is serviceable over a wider size range. A specific surface area of about 5000 sq. cm./g. (as determined by the Murex " Spekker " method) is suggested for the gamma alumina, but much finer material, e.g., surface area 15,000 sq. cm./g. is necessary in the case of alpha alumina.

Impurities of sulphur or chlorine in gamma alumina arising from unsatisfactory methods of preparation may result in etching or pitting of the specimen surface. There is no advantage in the use of deflocculants, which only reduce the rate of polishing.

Introduction

SATISFACTORY polishing is an essential preliminary to the microscopic examination of metallurgical specimens, and the most common method entails the use of a fine polishing powder on a piece of cloth. In this research, an attempt has been made to determine the physical characteristics necessary in a polishing powder, with special reference to alumina and its methods of preparation, with a view to obtaining the most suitable material for polishing purposes.

INFORMATION AVAILABLE IN THE LITERATURE

In most textbooks of metallography, *e.g.*, Greaves and Wrighton,[1] it is suggested that alumina suitable for polishing purposes can be obtained by the calcination of ammonium alum at high temperatures (1000–1100° C.). As an alternative to this method, Rowland and Romig[2] suggested the calcination of aluminium sulphate instead of ammonium alum : if serviceable, this method of preparation would have the advantage of obviating the formation of large clouds of ammonium sulphate fumes during the calcination, and might also prove to be a cheaper method of production. These authors recommend leaching with water to remove any sulphate present, and adjusting the *p*H of the aqueous suspension of the alumina with triethanolamine to a value of 7·0–7·6.

Recently it was suggested[3] that a superior alumina for polishing purposes was formed on the calcination of hydrated aluminium chloride, but no exact details of the time and temperature of treatment were given. Other authors, *e.g.*, J. L. Rodda,[4] have separated fractions of various sized gradings for polishing media.

These observations on the subject are, however, empirical in nature, and do not summarize the essential physical properties and conditions necessary for a good polishing medium.

Forms of Alumina

Some confusion has been caused by the nomenclature used by different authors in describing the hydrates of alumina, but Edwa rds[5] hasrecently pointed out that in the English system, these hydrates are named by the type of oxide which they give on calcination, so that diaspore, $Al_2O_3.H_2O$, is termed alpha in the English system, whereas the Americans call it a beta hydrate. In the English system, there are only two oxides now recognized, and they have the following physical characteristics[6] :

	Alpha alumina	Gamma alumina
Refractive index	1·760 (ϵ) 1·768 (ω)	1·696
Polarization colours	Greys	None (isotropic)
Specific gravity	3·91	3·47
Action of heat	m.p. 2050° C.	Reverts to alpha form at 1200° C. and over

These characteristics enable the two forms to be recognized readily in the laboratory—by their density, if a sufficient amount is available, and by microscopic or X-ray tests if small amounts only are at hand.

EXPERIMENTAL

Samples of alumina were obtained by calcination of various salts of aluminium under different conditions of time and temperature, and their size analyses checked both by the pipette method of sedimentation analysis, and by the direct measurement of their surface areas by the Murex optical method[7,8] ; other samples were also obtained from commercial sources, including polishing creams. A summary of the details of their preparation and polishing properties is contained in Table I. In some cases, the samples were elutriated to obtain different sized gradings for testing the importance of size of the medium particles on polishing.

Polishing tests were performed on mild-steel specimens (dia. $\frac{1}{2}$ in.) which were first ground on

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Dr. Rollason is Head of the Research Department, the late Dr. Sharratt was senior research chemist, and Mr. Roberts is assistant metallurgist, at Murex Welding Processes, Ltd., of Waltham Cross, Herts.

Table I—Physical Data and Polishing Properties of Various Samples of Alumina and other Metallic Oxides

Sample No. and Details	Through 300 Mesh B.S. Sieve	Sub-Sieve Analysis, per cent. less than						Spekker Surface Area (sq. cm./g.)	Density	Microscopy	Polishing
		30μ	20μ	10μ	5μ	2·5μ	1·5μ				
ALUMINA											
1 Commercial	100	100·0	98·7	75·0	40·8	15·1	5·9	5160	3·83	Granular particles of alpha alumina	Does not polish the surface but acts as a fine abrasive
1*a* Sample 1, elutriated − 5μ	100	100·0	100·0	94·5	92·3	55·3	35·6	16,900			Does not polish the surface but acts as a fine abrasive
2 From alum at 1100° C.	100	96·3	94·0	79·0	28·5	15·3	13·0	3690	3·47	Plate-shaped particles of gamma alumina*	Good surface in 4 min.
2*a* Sample 2, dispersed with $Na_4P_2O_7$ solution											Good surface in 5 min.
2*b* Sample 2, dispersed with triethanolamine aq. solution											Good surface in 5 min.
2c† Sample 2, elutriated + 17μ	100	100·0	98·6	56·9	25·4	14·8		3200			Scratches removed in 4 min., but surface slightly pitted
2*d*† Sample 2, elutriated − 17μ, + 5μ	100	100·0	100·0	84·3	43·8	17·4	6·8	5350			Good surface in 5 min.
2*e*† Sample 2, elutriated − 5μ	100	100·0	100·0	99·3	95·2	66·5	37·6	9800			Good surface in 5 min.
3 From alum at 1250° C. for 1 hr.	100	100·0	99·5	67·0	48·0	30·0	18·0	15,200	3·93	Small flakes of alpha alumina*	Good surface in 3 min.
3*a* From alum at 1250° C. for 5 hr.											Emery paper scratches removed in 3 min., but surface slightly scratched by alumina
4 From alum at 850° C.											Scratches removed in 3 min., but slightly pitted surface
4*a* From alum at 850° C., leached with water											Good surface in 5 min.
5 From aluminium sulphate at 1100° C. for 1 hr.	98·49							4590		Plate-shaped particles of gamma alumina	Scratches removed in 3 min., but slightly pitted surface
5*a* From aluminium sulphate at 1100° C. for 1 hr., leached with water											Good surface in 4 min.
6 From aluminium chloride at 500° C. for 1 hr.											Scratches removed in 7 min., but slightly etched surface
7 From aluminium chloride at 700° C. for 1 hr.	96·6	83·7	79·0	56·0	36·0	20·0	13·0	6600	3·18		Scratches removed in 5 min., but slightly etched surface
8 From commercial polishing creams, A	99·9	99·9	93·3	69·8	58·1	45·0	37·6	7350	3·23	Plate-shaped particles of gamma alumina	Good surface in 3 min.
9 From commercial polishing creams, B (3 hr.)	100	100·0	100·0	98·5	96·5	89·6	65·4	28,400	3·87	Form too fine to detect: mixed alpha and gamma aluminas	Good surface in 4 min.
NON-ALUMINA											
10 Cerium oxide	100	96·3	95·0	92·5	87·5	67·0	33·8	13,500	6·85	Flake-like. Elongated and aggregated	Good surface in 5 min.
11 Red iron oxide	100	88·0	78·5	66·5	58·5	57·5	39·5	27,000	5·13	Form too fine to detect. Aggregated	Emery paper scratches removed in 3 min., but iron oxide difficult to remove and conspicuous, and itself scratched the surface
12 Green chromous oxide	100	90·0	87·0	80·0	66·0	50·0	33·0	16,000	5·08	Small flakes	Emery paper scratches removed in 3 min., but slightly scratched surface

Notes: Size analyses were conducted by means of the pipette method, and surface areas by the Murex optical method using the Spekker photoelectric absorptiometer. For full details see references 7 and 8.

* Crystalline identity has been confirmed by X-ray powder photo, by courtesy of G. L. J. Bailey of B.N.F.M.R.A.

† The sample of gamma alumina elutriated was a further preparation of this substance, and not the one whose analysis is quoted under sample 2.

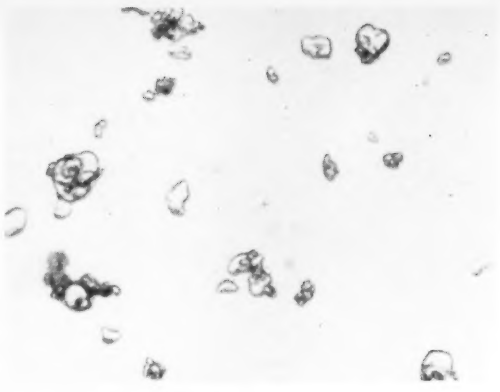

Fig. 1—Photomicrograph of globular particles of alpha alumina (sample 1) × 750

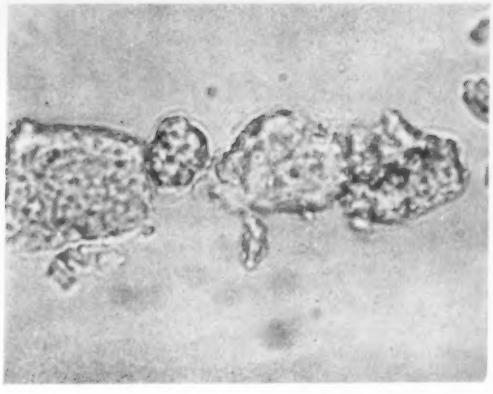

Fig. 2—Photomicrograph of plate-shaped particles of gamma alumina (sample 2) × 750

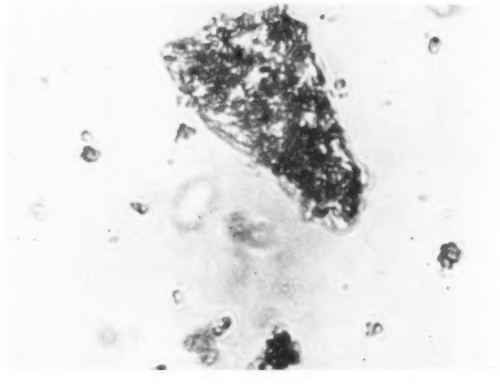

Fig. 3—Photomicrograph of alpha alumina obtained by calcination of alum at 1250° C. (sample 3) × 750

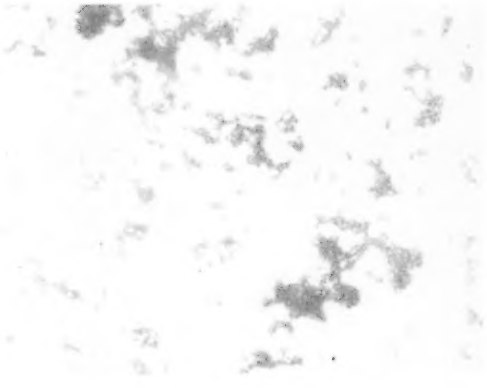

Fig. 4—Photomicrograph of small flakes of green chromous oxide, commercial sample (sample 12) × 750

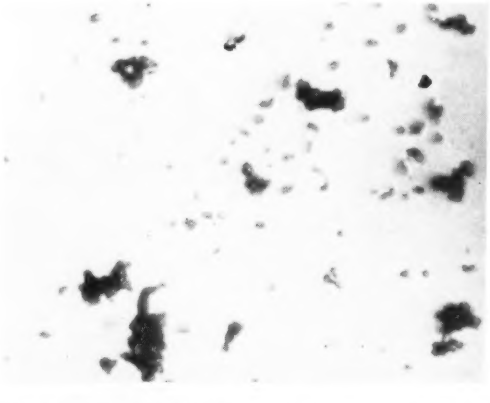

Fig. 5—Photomicrograph of elongated flakes of cerium oxide, commercial sample (sample 10) × 750

Fig. 6—Photomicrograph of aggregates of red iron oxide, commercial sample (sample 11) × 750

Notes : Size analyses were conducted by means of the pipette method, and surface areas by the Murex optical method using the Spekker photoelectric absorptiometer. For full details see references 7 and 8.

* Crystallime identity has been confirmed by X-ray powder photo, by courtesy of G. L. J. Bailey of B.N.F.M.R.A.

† The sample of gamma alumina elutriated was a further preparation of this substance, and not the one whose analysis is quoted under sample 2.

0000 emery paper, and then polished by hand on broadcloth, using a suspension of the polishing medium in distilled water, with or without additions, as stated in Table I. The time required to remove all the scratches was noted in each case and the specimens were examined under the microscope for signs of etching, etc. In addition, two of the alumina samples (Nos. 2 and 3) were tested on a specimen 1 in. square, and times of polishing were found to be 8 min. and 6 min. respectively, thus confirming the advantage of using flake-like alpha alumina.

Shapes of Particles of Polishing Media

The shapes of the particles of the various polishing media were compared by microscopic examination, and photomicrographs of some of the more important ones are shown in Figs. 1–6, all at magnifications of × 750 dia. All the substances were suspended in ethylene glycol, which is a very good dispersing agent with a low refractive index of 1·43. The following information was obtained from these Figures:

Fig. 1—These are clearly globular particles of alpha alumina, the high density confirming the microscopy of this commercial sample.

Fig. 2—Large plate-shaped particles of gamma alumina are clearly observable in this photomicrograph of alumina, obtained by treating alum at 1100° C. for 1 hr. The low density of this material confirms its identity as gamma alumina.

Fig. 3—Most of the alumina obtained by the calcination of alum for 1 hr. at 1250° C. was of very small size, but occasionally particles large enough to indicate their shape were found. It appears that the gamma alumina which is first formed in the calcination, recrystallizes *in situ* into the alpha form, and as a consequence of this recrystallization it retains the outward shape of the original gamma alumina, *i.e.*, plate shaped, as shown. This method of preparation has a marked effect on its behaviour as a polishing medium.

Fig. 4—Green chromous oxide consisting of small irregular flake-like particles. Some of these are just sufficiently transparent for the green colour to show through.

Fig. 5—The cerium oxide was very fine, but appeared to be flake-like rather than granular. It was aggregated. This material is much used for the polishing of glass.[9]

Fig. 6—Red iron oxide is so very fine that it is impossible to say for certain, from this photomicrograph, what is its form. It is highly aggregated.

Packing-Density Determinations on Some Media

Evidence regarding the form of particles in a powder can sometimes be obtained from packing-density determinations. According to Burmister,[10] the factors which influence packing density are: (i) degree of fineness; (ii) type of grading; (iii) range of particle sizes; and (iv) particle shape.

Amongst the samples of alumina under consideration, apart from elutriated samples, there is no major variable between (i), (ii), and (iii). Packing-density differences may therefore be attributed to variations in the shape factor, and here it may be noted that plate-shaped materials such as mica, are well known to give very low packing densities. The packing densities of various materials, together with their size data, are collected in Table II. From this table, it may be observed that the alpha alumina obtained by calcination of alum at 1250° C. for 1 hr. (sample 3) has a very much lower packing density than the granular form of this material (sample 1), and contains an even greater percentage of voids than does white titania, which, since it is exceptionally fine (median dia. = 0·30μ), has a normal type of size grading (straight line on log-probability plotting of particle-dia./cumulative-percentage-finer—*see* ref. 8, p. 36), and is the most uniform (least value of the standard deviation ρ) of all the materials listed, may be taken to represent the limiting effect of reduction in packing

Table II

PACKING-DENSITY DETERMINATIONS ON VARIOUS MATERIALS

Substance	Size Data			Packing Density	True Density	$\frac{\text{Packing D}}{\text{True D}}$	% Voids $= 10^{2}\left(1-\frac{PD}{TD}\right)$
	M, μ	ρ	A				
Commercial alumina (alpha, granular), sample 1	5·30	2·38	5160	1·390	3·83	0·363	63·7
Alumina from alum at 1250° C. for 1 hr. (alpha, flaky), sample 3	5·20	4·50	15,200	0·489	3·93	0·128	87·2
Alumina from alum at 1100° C. for 1 hr. (gamma, flaky), sample 2	4·80	2·67	3690	0·391	3·47	0·113	88·7
Titanium white	0·30	1·58	103,000	0·670	3·40	0·197	80·3
Mica	125·00	2·00	430	0·408	2·82	0·145	85·5
Green chromous oxide	2·75	5·93	16,000	1·520	5·08	0·299	70·1
Red iron oxide	2·65	10·40	27,000	0·995	5·13	0·194	80·6
Cerium oxide	1·90	(4)	13,500	1·410	6·85	0·206	79·4

A = "Spekker" surface area.
M = Median diameter = diameter for 50% by weight below size.
ρ = Standard deviation = $\frac{\text{Dia. for 84·1\% below}}{\text{Dia. for 50\% below}}$ or $\frac{\text{Dia. for 50\% below.}}{\text{Dia. for 15·9\% below.}}$

Size analyses were plotted[8] on log-probability paper, (particle-diameter/cumulative-percentage-finer). Most size analyses proved to be straight lines on this paper for all but the coarse extremity of the size plots, except in the case of cerium oxide: here the figure shown is that of a tangent to the curve at the 50% mark.

density likely to result from Burmister's factors (i)–(iii). Similarly, the low density of the gamma alumina (sample 2), which is only 11% of the true density, is primarily due to its plate-shaped particles. Hence, an independent proof of the plate-like nature of the particles of these forms of alumina may be deduced from their low packing densities.

The green chromous oxide shows only 70% voids, probably on account of the high value of the standard deviation ρ, which indicates wide size range (Burmister's factor (iii)), this figure being therefore what would be expected of a material consisting of small flakes and having a wide size range. The figure for cerium oxide shows a high voidage, consistent with a flake-like structure, but in view of the observed aggregation, it is difficult to say to what extent this may have influenced the result. For red iron oxide, there is again a high voidage, but the surface area is very high, and if the apparently high value of the standard deviation ρ, is due to aggregation, it is again difficult to say with certainty that this indicates a plate-like structure. A common feature of all the materials of value as polishing media (the granular material is unsuitable) is that they all have very low packing densities.

Microstructure of Polished and Etched Samples

To reveal differences in polishing and etching associated with powders of different particle shape, but not differing greatly in particle size, specimens of mild steel (dia. ½ in.), were polished with either flake-like gamma alumina (sample 2) or granular alpha alumina (sample 1*a*). The results showed that the specimen polished with flaky alumina had a superior polish, while the other specimen showed a number of scratches, and with somewhat prolonged polishing the pearlite was revealed in light relief. These specimens were then etched in 1% nital at room temperature (16° C.) and the following observations were made.

The surface of the specimen polished with granular alumina rapidly lost its brightness, which may have been due to the relative rates of attack on a polished, compared with an abraded, surface. Less than half the time was required for darkening the pearlite in this specimen as compared with that in the specimen polished with flaky alumina. The outline and resolution of the pearlite in the former specimen was, however, less clearly defined, suggestive that the distortion due to polishing had not been completely removed. Further, the ferrite grain boundaries were not revealed until the specimen polished with granular alumina had been etched for four times the duration of that used to obtain a similar etch on the other specimen. With "dead-soft" weld specimens this difference was even greater, about fifteen times the duration of etching being required to reveal the grain boundaries on a specimen polished with granular alumina as compared with a similar specimen polished with flaky alumina.

One explanation of these results is that the pressure developed beneath flaky polishing powder is lower than that produced with granular powder. The high local pressure results in the metal being cold worked to a greater depth, but the degree of breakdown of the crystalline structure is likely to be less than that produced by the flaky powder, which may be considered as forming a shallow flow layer with negligible crystalline form, *i.e.*, amorphous.

GENERAL DISCUSSION OF RESULTS

Effect of Shape of Particles

The outstanding fact relating to the materials found serviceable as polishing agents is that the *shape* of the particles is invariably plate-like. Of the various forms of alumina examined, the gamma form obtained by the calcination of alum at 1100° C. for one hour (sample 2) is extremely valuable, and, as discussed later, may be used over a wide size range. The alpha form, on the other hand, is unsuitable in the granular form normally obtained commercially (sample 1), for it scratches the specimen under test; but if this same alpha alumina is prepared by calcination it consists of flat rectangular plates,[11] and a very different polishing result is obtained. The calcined alpha alumina (sample 3) is, in fact, the best polishing medium of all, for it is more rapid in its action than the softer gamma alumina—alpha alumina is said[12] to have a Mohs' hardness of 9, whereas gamma alumina has a hardness of 8.

Overheating of the alpha alumina in this calcination is, however, undesirable. Thus, when the heating at 1250° C. was prolonged to five hours, the product obtained caused a little scratching: this may be due to local sintering which could cause the formation of larger particles more liable to produce scratches, and may, conceivably, alter the particle shape to some extent.

The commercial polishing cream *B* consists mainly of alpha alumina, but the presence of some gamma alumina may be due to its having been obtained by heat-treatment insufficient to cause the complete conversion of the material to the alpha form. This is a very fine form of alumina, from which coarse particles have been removed by sedimentation. Kenyon[12] only uses alpha alumina for rough polishing, and "Gamal" (gamma alumina ?) for the final polish.

The flake-like shape of the cerium oxide, also found to give a good surface (sample 10), confirms the importance of the shape factor necessary for a good polishing medium.

HYPOTHESIS OF THE MECHANISM OF METALLOGRAPHIC POLISHING

It is possible to suggest a hypothesis of the mechanism of metallographic polishing based on the shape of the particles. The main characteristic of a mechanically polished metal surface is the presence of the Beilby layer[13] of amorphous metal to a depth of some 20–30 atoms. Such flow layers are likely to arise from the frictional heat effects between plate-like abrasives and the "hillocks" of the ground metal surface. The thickness of the flow layer is probably affected by the lubrication effects of the water, but is likely to be much thinner than the worked layer produced with the "cutting" action of granular powder. Removal of the flow layer with etching leaves a surface relatively free from the distortion effect of the polish, although previous grinding effects may be evident.

With this mechanism of polishing, the size of the

abrasive "plates" is not likely to be critical, and this is confirmed by the results already obtained. When polishing with a granular abrasive, however, the relatively sharp edges produce a cutting action analagous to that of a lathe tool, and tiny grooves remain as scratches, with distortion beneath (*see* Fig. 7). If the granules are sufficiently small, the scratches become less significant when viewed under the microscope. Such a polishing powder is likely to be fairly rapid in its cutting action, but for equal degrees of polish its size is likely to be far more critical than that of a platy powder. The deep-worked layer, produced with a considerable degree of crystallinity, will require prolonged etching for removal.

MISCELLANEOUS FACTORS AFFECTING POLISHING RESULTS

Effect of Impurities in the Alumina

When the alum was calcined at a lower temperature, *e.g.*, 850° C. (sample 4), the surface produced was pitted, because of an impurity of sulphate. Removal of this sulphate by leaching with water (sample 4*a*) gave a powder serviceable for polishing, but slower in its action than the alumina prepared by direct calcination at 1100° C. (sample 2). Similarly, when aluminium sulphate was used as the starting material (sample 5), calcination at 1100° C. was insufficient to remove all the sulphate, and a pitted surface was again produced. This result is in accord with Rowland and Romig's finding,[2] who also observed sulphate impurity in alumina prepared from aluminium sulphate at 1000–1100° C. for 2–5 hr., and who treated their product by leaching with water. There appears to be no advantage in these alternative methods of preparing alumina, and as they involve leaching with water to remove undesirable impurities, it would appear better to adopt ammonium alum as the source of the material, and to use the higher temperature of 1100° (sample 2) or 1250° C. (sample 3), since these heat-treatments produce materials ready for use as polishing media without further treatment such as leaching.

Yet another possible source of the alumina is hydrated aluminium chloride,[3] but on testing, this was found to result in products containing some chlorine, which caused etching of the specimen surface (*see* tests 6 and 7, Table I).

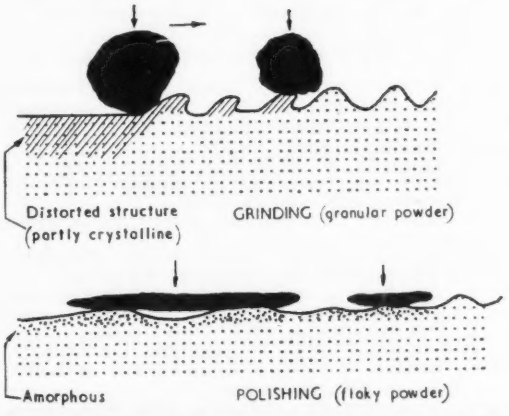

Fig. 7—Diagrammatic representation of the grinding and polishing mechanisms

Rate of Polishing and Time of Calcination of Alum

R. W. Urie and A. W. Wylie[9] found that when rare earth oxides (chiefly cerium oxide) were prepared from hydroxides by calcination, for use as materials for polishing glass, the rate of polishing varied with the time of heating at 1150° C. It was therefore decided to test the effect of time of ignition of the alum at 1100° C., on the rate of polishing of the alumina obtained. No difference in rate of polishing could be found between samples which had been heated for 1 or 5 hr. at 1100° C., but if only ½ hr. ignition was given, some sulphate still remained in the sample and caused etching of the specimen under test. It is therefore concluded that in the preparation of gamma alumina for polishing purposes, the time of ignition at 1100° C. should be at least 1 hr., but that no further advantage accrues from prolonging the heating at this temperature. (These times are quoted for experiments using about 500 g. of alum).

Effect of Size of Particles of Medium

The elutriation of sample 2 (gamma alumina) did not succeed entirely in separating the material into clear-cut size fractions, as with previous observations on the Schone elutriator by Stairmand.[14] However, it did produce samples of different sized gradings, *viz.*,

Sample 2*c*—consisting mainly of coarser material with a median particle dia. of about 10μ; and surface area 3200 sq. cm./g.

Sample 2*d*—containing no material greater than 20μ in size, and having a median dia. of about 5μ, and surface area 5350 sq. cm./g.

Sample 2*e*—containing little material over 5μ size and having a median dia. of about 2μ, and surface area 9800 sq. cm./g.

Comparison of the results obtained on polishing with these samples shows that none of them are of any greater value than the original sample No. 2, *i.e.*, as calcined and before elutriation. On the other hand, excessive coarser material was found to be disadvantageous (sample 2*c*), whilst finer fractions required slightly longer in polishing (sample 2*e*). This latter finding is confirmed by the time required for polishing with cerium oxide (sample 10), which has about the same size analysis as the elutriated -5μ fraction (sample 2*e*). It is concluded, therefore, that samples of gamma alumina are suitable for polishing purposes provided that they contain no material of size greater than about 17μ, the preferred surface area being about 5000 sq. cm./g.; apart from these limitations, the exact size-grading of the material into uniform and precise size fractions is unnecessary, whilst too fine material may suffer from the disadvantage of being too slow in its action.

With alpha alumina, however, it would appear advisable to use a finer-sized material. The only two samples of alpha alumina found successful for polishing purposes were sample 3, of surface area 15,000 sq. cm./g., and the commercial cream *B*, of surface area 28,000 sq. cm./g., which consisted mainly of alpha alumina. Elutriation of the granular form of alpha alumina to give sample 1*a*, of surface area

16,900 sq. cm./g., gave a material which still caused some scratching of the surface (Table I).

Effect of *p*H of Suspension

The alumina obtained from the calcination of alum at 1100° C. (sample 2) was suspended in liquids of *p*H varying from 4 to 10, the adjustments being made by slight additions of either ammonia or acetic acid, as required.

It was found that, to avoid etching, the *p*H of the suspension should be neutral or faintly alkaline, *e.g.*, between *p*H 7·0 and 9·0 ; and that within the range *p*H 5–10, the rate of polishing did not vary.

Effect of Deflocculants

Dilute suspensions of alumina settle rather rapidly, particularly if the *p*H is near the neutral point, settling being less rapid as the *p*H moves away from the neutral point in either direction. Hence it was thought that the addition of deflocculants might have some advantage, but tests 2*a* and 2*b*, using sodium pyrophosphate or triethanolamine, respectively, (4 g. alumina, 0·28 g. deflocculant, in 200 c.c. distilled water), showed that both caused a slight reduction in the rate of polishing. When Dispersol T was used in similar proportion, the rate of polishing was still slower, eight minutes being required.

It is therefore concluded that the use of deflocculants possesses no marked advantage, and tends to reduce the rate of polishing.

The Use of Metallic Oxides Other Than Alumina

Three metallic oxides other than alumina have been tested for polishing mild steel, *viz.*, cerium oxide, a material used extensively for polishing lenses,[9] *etc.*, in the glass industry, green chromous oxide, and red iron oxide. Data regarding their size analyses and polishing values are included in Table I.

CONCLUSIONS

The following conclusions are drawn from this investigation into the factors effective in determining the value of a polishing medium for mild-steel specimens :

(1) Shape of the particles is the most important factor ; plate-shaped materials are good agents but granular materials should be avoided, and a hypothesis of the mechanism of polishing based on the shape of the particles has been suggested. Flow layers may arise from the frictional effects between plate-like abrasives and the " hillocks " of the ground metal surface, whereas granular materials produce a cutting action analagous to that of a lathe tool, so that tiny grooves remain as scratches, together with a worked layer beneath.

(2) Alumina is preferred of a group of several oxides for polishing purposes. Both alpha and gamma forms of alumina exist, and in the plate-like form, either may be used, subject to the following conditions :

Alpha Alumina is obtainable as a fine material suitable for polishing if prepared from alum by calcination at 1250° C. for 1 hr., when material of median size 5μ and surface area 15,000 sq. cm. is obtained. Longer heating at this temperature must be avoided as local sintering to give larger particles (which cause scratching) may occur. Normal commercial samples of alpha alumina are granular, and are unsuitable for polishing unless exceptionally fine.

Gamma Alumina is the material obtained when the textbook method of preparation (alum calcined at 1100° C. for 1 hr.), is used. It is not quite so rapid in its polishing action as alpha alumina, but has the advantage that it gives a scratch-free surface over a wider size range than does the alpha material. It may therefore be observed that alpha alumina prepared by calcination is a more rapid polishing material than is gamma alumina, but it is questionable whether this advantage is sufficient to offset the effort of obtaining under laboratory conditions the high temperature required and the greater size limitations demanded for the alpha form.

(3) Acid impurities are liable to cause etching ; they are most easily avoided by using ammonium alum as source of the gamma alumina, and calcination at 1100° C. for at least 1 hr. Prolonged heating at this temperature appears to offer no advantage.

(4) The *p*H of the suspension should be between 7 and 9, to avoid etching.

(5) Gamma alumina coarser than 17μ in size or less than 3500 sq. cm./g. surface area, is undesirable, but apart from this limitation, no particular advantage accrues from the use of material of specific size gradings, and, if too fine, the speed of polishing may be reduced ; gamma alumina of surface area 4000–5000 sq. cm./g. is preferred. If alpha alumina is used, however, a finer size surface area, 15,000 sq. cm./g., is suggested.

(6) There is no advantage in the use of deflocculants in the alumina suspension, and these substances may reduce the rate of polishing.

Acknowledgments

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References

1. R. H. Greaves and H. Wrighton : " Practical Microscopical Metallography," p. 9. London, 1939 : Chapman and Hall.
2. D. H. Rowland and O. E. Romig : *American Society for Metals*, 1943, vol. 31, p. 980.
3. Anon., " Aluminium from Clays," *Chemical and Engineering News*, 1947 (June), vol. 25, p. 1590.
4. J. L. Rodda : *Transactions of the American Institute of Mining and Metallurgical Engineers*, 1932, vol. 99, p. 149.
5. J. D. Edwards : *Chemistry and Industry*, 1947, p. 230.
6. J. D. Edwards, F. C. Frary, and Z. Jeffries : " The Aluminium Industry," vol. I, " Aluminium and its Production," p. 164. New York, 1930 : McGraw-Hill Book Co., Inc.
7. E. Sharratt, E. H. S. van Someren, and E. C. Rollason : *Journal of the Society of Chemical Industry*, 1945, vol. 64, p. 73.
8. E. Sharratt : *Transactions of the Ceramic Society*, 1948, vol. 47, p. 22.
9. R. W. Urie and A. W. Wylie : *Journal of the Society of Chemical Industry*, 1947, vol. 66, p. 433.
10. D. M. Burmister : *Proceedings of the American Society for Testing Materials*, 1938, vol. 38, p. 587.
11. B. A. Scott : *Nature*, 1948, vol. 161, p. 358.
12. C. P. Kenyon : *Metal Progress*, 1947, vol. 51, p. 260.
13. G. Beilby : " Aggregation and Flow of Solids," London, 1921, Macmillan.
14. C. J. Stairmand : " Particle Size Symposium," London, 1947, p. 58. The Institution of Chemical Engineering.

Change of Electrical Resistance during the Strain Ageing of Iron

By A. H. Cottrell, Ph.D., and A. T. Churchman, B.Sc.

SYNOPSIS

It has been suggested recently that the yield point and strain ageing of α-iron are due to the migration of carbon and nitrogen atoms to dislocations. One consequence of the theory is that the electrical resistance should decrease by a small and roughly predictable amount on strain ageing. This effect is examined experimentally on soft iron wire and is shown to exist. The rate of change of the resistance is increased by increasing the temperature of ageing, corresponding to an activation energy of 16,000–22,000 cal./mol. This agrees with the view that the change is caused by the migration of carbon and nitrogen, since the activation energy for the diffusion of these elements in α-iron is about 20,000 cal./mol. The magnitude of the resistance change, approximately 0·2%, is increased by increasing the amount of cold work. Further working and ageing cause the change to be repeated. These effects are shown to be consistent with the dislocation theory and can be used to estimate the density of dislocations in cold-worked iron.

Introduction

RECENT work suggests that the elastic limit of a crystalline material is determined by the stress required to enable dislocations, trapped within the crystals, to escape by overcoming the energy barriers of their traps and to move at high speeds through the lattice.[1] These energy barriers are caused by the interactions of the dislocations, by means of their stress fields, with nearby irregularities which are also sources of internal stress. The yield point of iron has been discussed from this point of view in terms of a special form of trap involving the segregation of solute atoms round dislocations,[2] as follows. In a positive Taylor[3] or edge[4] dislocation, the region of the lattice above the dislocation is compressed and that below is expanded. Suppose that solute atoms near the dislocation rearrange themselves so that those which expand the lattice go to the expanded region, and *vice versa*. This special segregation will cause the natural (*i.e.*, stress-free) lattice parameter at any point near the dislocation to approach more closely to the actual local value of the parameter. Thus the local internal stresses will be partly relieved and the strain energy will be reduced, although the *strain* field of the dislocation will be hardly altered. It is to be expected, therefore, that a dislocation will gather round itself an 'atmosphere' of solute atoms and that those atoms which strongly distort the lattice should be especially involved. An additional interaction with dislocations is possible in the case of solute atoms which distort the lattice non-symmetrically (*e.g.*, carbon and nitrogen in α-iron), because these atoms can relieve shear stresses round dislocations, as well as hydrostatic stresses.[5, 6]

When an external stress is applied, each dislocation will try to move and leave its atmosphere behind. The strain energy is thereby increased so that a force is exerted holding each dislocation to its atmosphere. Only when the applied stress is large enough to overcome this anchoring force can a dislocation break away and move freely through the lattice. However, once the dislocations are free, the stress that has freed them is more than is necessary merely to keep them in motion, and they must therefore accelerate. Rapid flow under smaller stresses is possible, and the material should show a sharp upper yield point followed by plastic flow at a lower yield point.

The resemblance which this bears to the well-known behaviour of soft iron and steel becomes significant when it is remembered that the sharp yield point of iron is associated with the presence of small amounts of carbon or nitrogen, or both, in the metal[7–9] and that these elements strongly distort the α-iron lattice when dissolved interstitially.[10, 11] Accordingly, the theory has been developed[2, 5, 6] that the yield point of iron is caused by the segregation of carbon and nitrogen atoms to dislocations in the metal. The removal of the sharp yield point by plastic over-strain and its return on strain age-hardening can be explained by the theory as follows. If a specimen has been loaded beyond its yield point and then unloaded, it contains free dislocations and will not show a yield point on immediate reloading. However, if it is aged before reloading, the solute atoms are given time to migrate to the dislocations to form new atmospheres, and the yield point returns. Nabarro[5] has shown that the activation energy for strain ageing is the same as for the diffusion of carbon in α-iron and, more recently, it has been shown[6] that the observed rate of strain ageing can be accounted for if it is assumed that about 0·003 wt.-% of carbon is in solution.

Professor Casimir (Eindhoven) pointed out to one of the authors (A.H.C.) in discussion that, on this theory, one would expect the electrical resistance of iron to decrease by a small amount on strain ageing. Carbon and nitrogen atoms in solution in α-iron scatter electrons very effectively; the electrical resistance is thus highly sensitive to the amounts of these elements in solution, and it has been shown that a decrease of 0·01 wt.-% in solution causes a decrease in resistance of 2·5% in the case of carbon[11, 12] and 3·5% in the case of nitrogen.[11] Although the transfer of atoms of these elements from random solution to

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Dr. Cottrell and Mr. Churchman are at the Department of Metallurgy, University of Birmingham.

dislocations cannot be so effective in reducing the resistance as would removing them completely, it is nevertheless reasonable to expect a change of comparable, though smaller, magnitude.

The purpose of this paper is to report an investigation of the electrical resistance of iron during strain ageing, which was made with the object of testing the above prediction.

APPARATUS AND TECHNIQUE

Before planning the experiments, an estimate was made of the expected change of resistance. It has been argued[6] that a dislocation atmosphere in iron containing carbon probably takes the form of a single line of carbon atoms, lying parallel and very close to the dislocation, with one carbon atom for every atomic spacing along the dislocation line. From published estimates,[1, 3, 13–15] it seems certain that the density of dislocations in cold-worked iron is, at least, 10^{10} cm.$^{-2}$, and that it can be as high as 10^{12} cm.$^{-2}$. Taking 10^{10} as a lower limit, at least 0·0001 wt.-% of carbon ought to be transferred to the dislocations from solution during strain ageing. An upper limit is fixed by the total amount of carbon, which, according to Yensen,[16] does not exceed 0·006% at room temperature. Using as a rough guide the relation connecting carbon content and resistance, given in the previous section, a resistance change of between 0·025% and 1·5% was expected. Thus an experimental method sensitive to a resistance change of 1 : 10,000 was required.

A potentiometric method was used which consisted of measuring and comparing the potential falls across standard and test-specimen wires, connected in series, through which a small current was passed only when measurements were being taken. The potential falls were measured on a Tinsley potentiometer, reading to 0·5 mV., and a Tinsley moving-coil galvanometer arranged to give a deflection of 10 cm. at 1 m. for a potential fall of 0·5 mV. Changes could be detected to 0·0025 mV. so that, by maintaining a 50-mV. fall across each wire, the required sensitivity was possible. Mercury switches were used for bringing the wires in turn into the potentiometer circuit and also for reversing the potential and current. By averaging two successive readings with opposite potential and current, parasitic E.M.F's could be eliminated from the measurements. It was possible to make a complete measurement of both the standard specimen and the test-specimen within 30 sec., but to minimize the heating effect of the current passing during this period, a current of not more than 0·1 amp. was used. This restriction made it necessary to use a 12-ft. length of wire to obtain a potential fall of 50 mV.

Since a temperature change of 1° C. alters the resistance of iron by about 0·5%, it was imperative to prevent the temperatures of the standard and test wires from changing *relative to each other* during the course of an experiment, and the following precautions were therefore taken :

(1) The standard wire used was identical with the test wire in every respect except that it was fully aged before use.

(2) The wires were wound into 3-in. dia. coils on two identical, skeleton-like formers, of low thermal capacities and made from strips of ' Tufnol ' board. The standard wire was aged after mounting on its former.

(3) During the entire course of an ageing experiment, the wires were immersed side by side, and were left undisturbed in a well-lagged and continuously stirred oil bath. It was thus arranged that the responses of the two specimens to a change in the temperature of the oil bath were as similar as possible. The temperature of the bath was maintained constant to within $\pm 0{\cdot}5°$ C. by means of a bimetallic thermostat.

(4) Preliminary experiments were made with fully aged specimens which showed that, by using the above arrangements and neglecting measurements taken within the first 5 min. of immersion and at the times when the thermostat heater was operating, no detectable variations in the relative resistances of the wires occurred. By taking the same precautions in an actual experiment, it was thus ensured that all observed changes were in fact due to the ageing process itself.

Experiments were made on specimens from a coil of soft iron wire, 0·064 in. initial dia., having the following analysis :

Carbon	0·07%	Manganese	0·47%
Sulphur	0·026%	Silicon	<0·01%
Phosphorus	0·035%	Nitrogen	0·004%

The wire was annealed at 900° C. for 30 min., air-cooled to room temperature, descaled, and then stored at room temperature for at least one month before use.

To prepare a specimen for ageing, a 12-ft. length of wire was strained by cold drawing at a slow speed, using a tallow lubricant ; the wire was then wound on its former, and the electrical connections were made by fastening its ends to copper grips attached to the frame of the former. The time between the end of drawing and the immersion of the mounted specimen, with its standard, in the oil bath, was fixed at 10 min.

CHANGE OF RESISTANCE ON AGEING

Figure 1 shows two examples of the resistance change observed during ageing at 21° C. on specimens which were reduced 23% in diameter by drawing. Because it was impossible to obtain reliable measure-

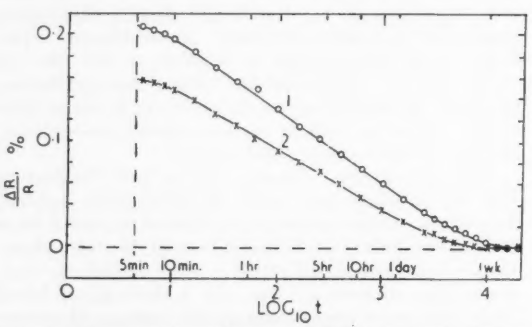


Fig. 1—Change of resistance ($\Delta R/R$) of cold-drawn iron wire (23% reduction) with time (t, in minutes) of immersion in the ageing bath at 21° C.

ments within the first 5 min. of immersion, the change has been expressed in terms of the resistance at the end of ageing, instead of the beginning, as a reference point. Thus $\Delta R/R$ in Fig. 1 is defined as $(R_t - R_\infty)/R_\infty$, where :

$$R_t = \frac{\text{P.D. across specimen}}{\text{P.D. across standard}} \times \text{resistance of standard}$$

measured at time t, and R_∞ is the same quantity at the end of the change. The results show that the resistance decreases by about 0·2% over the period in the oil bath, and that the change is fairly reproducible in time and magnitude ; in general, the reproducibility in repeated experiments was better than Fig. 1 suggests. It is to be noted, remembering that the time scale is logarithmic, that the change is fastest in the early stages of ageing and that it slows down as it develops. Thus, it is one-quarter complete in 30 min., and one-half complete in 3 hr., but it does not end within a week. This is qualitatively consistent with the known characteristics of strain age-hardening[17] and also with the dislocation theory. It has been shown[6] from the theory that strain ageing should start without an induction period and follow a $t^{\frac{1}{2}}$ law in its early stages, after which it should deviate from this law to become much slower.

EFFECT OF AGEING TEMPERATURE

Having established the existence of a resistance change on ageing, further information was sought by examining the effect of ageing at different temperatures. Experiments were made at temperatures of 21°, 31°, and 41° C. on specimens reduced 23% by drawing. The reproducibility at 31° and at 41° C. was about the same as at 21° C., but at higher temperatures the change was too fast to be followed accurately. Two experiments were made at each temperature, and their results were combined to give the mean ageing curves in Fig. 2. These curves show that the resistance change occurs more rapidly at higher temperatures, the rate being multiplied two to three times for a 10° rise, and that the total magnitude of the change, from 5 min. onwards, is reduced.

If the resistance change is caused by the migration of carbon and nitrogen atoms from random solution in the lattice, then its rate should be determined by their rate of migration, and it should possess the same activation energy as for the diffusion of these elements in α-iron. Studies of the elastic relaxation of iron

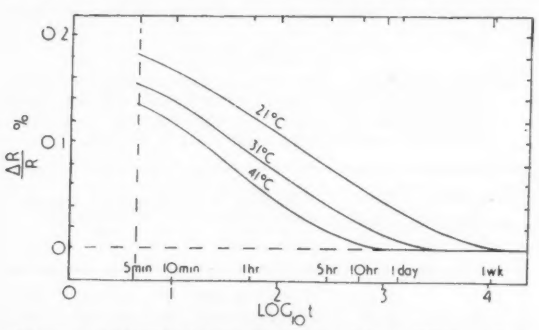


Fig. 2—Resistance/ageing curves of iron (23% reduction) at different temperatures

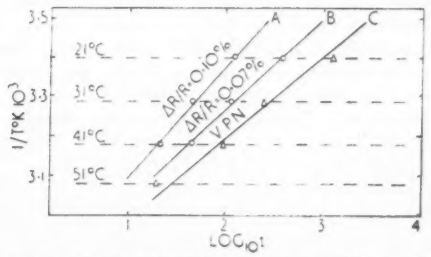


Fig. 3—Activation energies associated with resistance ageing and strain age-hardening. Lines *A*, *B*, and *C* correspond to activation energies of 17,000, 20,000, and 22,000 cal./mol., respectively

have given the activation energy for diffusion of nitrogen as 16,400 cal./mol. (Snoek[10]) and 20,000 cal./mol. (Kê[18]) ; Polder[19] has analysed Snoek's results[10] to determine the diffusion coefficient of carbon as :

$$D = 5{\cdot}2 \times 10^{-4} \times e^{-18{,}000/RT} \text{ cm.}^2\text{sec.}^{-1}.$$

Before the results of Fig. 2 are compared with these activation energies, a small correction needs to be made to allow for the fact that the specific resistance of iron is slightly different at the different temperatures. The temperature coefficient of resistance of the wire was found to be $3{\cdot}8 \times 10^{-3}$, so that the ratios of the resistances of a wire at 21°, 31°, and 41° C. respectively, were 1 : 1·04 : 1·07. Accordingly, the values of $\Delta R/R$ for the three temperatures were multiplied by these factors in order to bring the three resistance changes on to a common basis as fractions of the resistance of the fully aged wire at 21° C. To determine the activation energy, the standard method was used in which the times to reach a certain stage in the change are plotted logarithmically against the reciprocal of the absolute temperature ; if an activation energy exists, this should give a straight line, from the slope of which the activation energy can be found.

Figure 3, lines *A* and *B*, shows the results obtained at the stages $\Delta R/R = 0{\cdot}10\%$ and $0{\cdot}07\%$, respectively ; the slopes of these lines correspond to activation energies of 17,000 cal./mol. and 20,000 cal./mol. There was a general tendency for the activation energy to increase as the change progressed, the early stages being characterized by a value of 16,000–18,000 cal./mol., and the later stages by a value of 20,000–22,000 cal./mol. This may mean that more than one solute is taking part in the process, the species with the lower activation energy for migration being transferred first, which is consistent with the idea that both carbon and nitrogen take part ; but it may be due to experimental error, since the apparatus was being used at the limit of its sensitivity. The best mean activation energy for the resistance change lies in the range 18,000–20,000 cal./mol., which agrees with the view that the process depends on the migration of carbon or nitrogen.

MAGNITUDE OF THE CHANGE

Figure 2 shows that the total changes observed after the first 5 min. of immersion are 0·185%, 0·155%, and 0·135%, respectively, at the temperatures of 21°, 31°, and 41° C. Correction for the different specific resistances of iron at the three

temperatures brings these results a little closer together, *i.e.*, 0·185%, 0·161%, and 0·145%. However, a more important effect arises from the fact that the amounts of ageing which go unrecorded during the first 5 min. are different at the three temperatures. Taking the activation energy to be 18,000 cal./mol., 5 min. at 31° and 41° C. are equivalent respectively to 14 and 34 min. at 21° C. Thus, the amount of change observed at 31° C., for example, should be the same as that observed at 21° C. starting from 14 min. after immersion. The changes at 31° and 41° C. predicted in this way from the results at 21° C. are 0·16% and 0·14%, which are almost the same as the observed changes at these temperatures.

The results thus confirm that the total change at 21° C. after the first 5 min. is about 0·18%. To obtain the complete change, an addition must be made of the change which occurred during the time before measurements were started (10 min. at room temperature between drawing and immersion, and 5 min. of immersion at 21° C.). An estimate based on the observed initial rate of ageing shows this to be about 0·03%, thus giving the complete change as 0·21%.

It is interesting to consider the changes of carbon or nitrogen content and the density of dislocations that correspond to this value. Since transferring carbon atoms to dislocations cannot be more effective in reducing the resistance than removing them completely, a lower limit to the amount transferred can be obtained by taking the resistance change accompanying a change of 0·01 wt.-% of carbon to be 2·5%.[11, 12] This gives a transfer of 0·0008 wt.-% of carbon on strain ageing, which is below proposed values for the upper limit of residual carbon in solution.[11, 16] A similar result holds if nitrogen is taken to be the element concerned.

While the present investigation was in progress, results of a theoretical analysis by Bhatia[20] became available which enabled an upper limit to the amount of carbon transferred to be calculated. The authors are grateful to Dr. Bhatia for communicating his results to them before their publication. In his work the scattering of electrons by carbon atoms was analysed by wave-mechanical methods to find the change in resistance caused by assembling a given number of carbon atoms from random solution into rows, corresponding to their segregation to dislocations. Carbon atoms in the rows make only a small contribution to the resistance when the rows are parallel to the direction of the current flow, and even with a random arrangement of the orientations of the rows this contribution to the resistance is less than when the carbon atoms are distributed at random. The calculations show that transferring 0·01 wt.-% of carbon into randomly arranged dislocations ought to cause the resistance to decrease by 0·3%. Applying this to the present results gives the transfer of carbon as 0·007%, which is rather high when compared with the expected residual solubility of carbon.[11, 16] However, this figure is probably an upper limit, since it is based only on the effect of ordering the solute atoms into rows; an additional contribution to the resistance change will result from the fact that, when a carbon atom moves into a dislocation, the lattice strain round the original site of that atom is removed without a corresponding increase appearing in the strain near the dislocation.

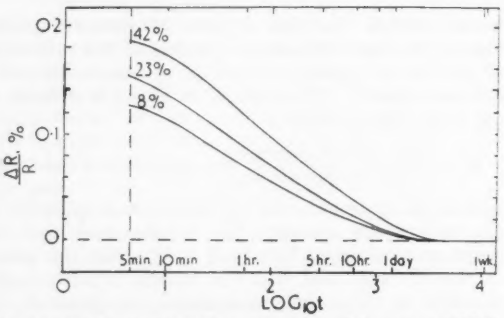


Fig. 4—Resistance ageing curves of iron at 31° C., after different reductions by cold drawing

The amount of carbon transferred is expected, therefore, to lie within the limits 0·0008–0·007%. Assuming that each dislocation requires one solute atom per atom plane along its length, the densities of dislocations corresponding to these amounts of carbon are 8×10^{10} cm.$^{-2}$ and 7×10^{11} cm.$^{-2}$, respectively. Similar results are obtained if the estimate is based on the transfer of nitrogen instead of carbon. A dislocation density of the order of 10^{11} cm.$^{-2}$ is very reasonable for a sample of iron which has been subjected to a moderate degree of cold work (23% reduction).

EFFECT OF AMOUNT OF DEFORMATION

A series of experiments was made to examine the effects of varying the amount of cold work on the change of resistance during ageing. Reductions of 8% and 42% by drawing were used, and three ageing curves were taken at 31° C. for each reduction. Mean ageing curves derived from these results are given in Fig. 4, together with the curve, already determined, for 23% reduction.

It will be seen that the amount of cold work has little effect on the time of ageing. However, the magnitude of the resistance change is affected, and the total changes, after correction, are as given in the first two columns of Table I. These results can be interpreted in terms of an increase, with increasing deformation, in the number of centres at which the process occurs which causes strain ageing, without

Table I

ESTIMATES OF CHANGES OF CARBON AND DENSITIES OF DISLOCATIONS FROM THE RESISTANCE CHANGES OBSERVED FOR DIFFERENT AMOUNTS OF COLD WORK

Reduction by Drawing, %	Total Change of Resistance, %	Estimated Amount of Carbon Transferred to Dislocations, %	Estimated Density of Dislocations, cm.$^{-2}$
8	0·18	0·0007–0·006	7×10^{10}–6×10^{11}
23	0·21	0·0008–0·007	8×10^{10}–7×10^{11}
42	0·24	0·0009–0·008	9×10^{10}–8×10^{11}

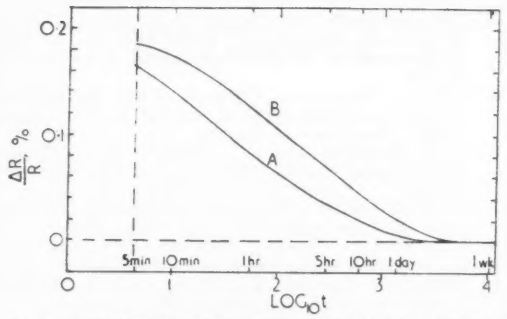


Fig. 5—Resistance ageing curves at 31° C. Curve *A*, after 23% reduction; curve *B*, after complete ageing and a further 23% reduction

any change occurring in the individual characteristics of the centres. This is in accord with an increase in the density of dislocations. Each dislocation collects carbon atoms at a characteristic rate, and thus the time for strain ageing is unaffected by the amount of deformation. The number of dislocations is altered, however, which leads to a corresponding alteration in the magnitude of the resistance change. The densities of dislocations and amounts of carbon transferred for each of the three reductions, as derived by the method used in the previous section, are given in Table I.

While it is a widely held view that work hardening is due to an increase in the density of dislocations, the details are not yet clear. In his original formulation of the theory, Taylor[3] proposed that a regular or quasi-regular array of dislocations would be built up, from which it can be deduced that the hardness should increase inversely as the distance between neighbouring dislocations. In the present case, this would mean that the square of the hardness should be proportional to the resistance change. Other views of work hardening, which have been developed more recently[13, 21] suggest that the hardness might increase linearly with the density of dislocations. The Vickers pyramid hardnesses of the freshly worked wires deformed by 8%, 23%, and 42%, were determined as 140, 170, and 180, respectively; the ratio of these values is 1 : 1·21 : 1·29, and the ratio of their squares is 1 : 1·47 : 1·65. The corresponding ratio of the resistance changes in Table I is 1 : 1·17 : 1·33, which agrees rather better with the later theories

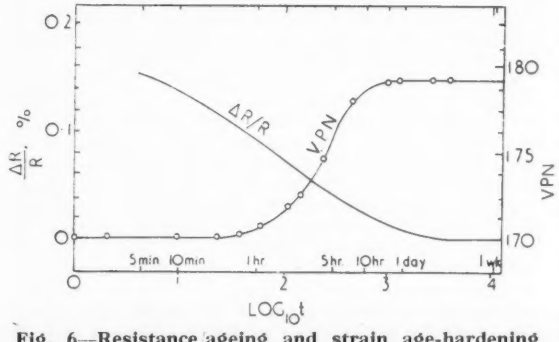


Fig. 6—Resistance ageing and strain age-hardening curves at 31° C. of iron reduced 23% by drawing

of hardening than with the original theory. Although these results are hardly accurate enough to form a definite conclusion on this point, they suggest that more accurate resistance measurements might allow a critical test of the theories to be made.

REPEATED AGEING EXPERIMENTS

Some experiments were made to examine the ageing behaviour of wires which were strained a second time after being subjected to a complete strain-ageing treatment. The wires were first reduced 23% by drawing and were aged at 31° C. in the usual manner. The resistance curves were determined during ageing, and the wires were held at 31° C. for 3 weeks to ensure that ageing was completed. A further reduction of 23% was then given (making a total reduction of 42% on the original diameter), and the ageing curves were again determined. The mean results for two tests given in Fig. 5 show that the second treatment produces a resistance change similar to the first. The time of ageing is about the same, although the total change is larger. The ageing curve in the second treatment is, in fact, almost identical with that for wires reduced directly by 42%, without an intermediate ageing treatment (*cf.* Fig. 4), showing that the ageing treatment after the first stage of reduction has virtually no effect on the second curve.

These results can be interpreted as follows. At the end of the first ageing treatment the dislocations are all anchored by their lines of carbon or nitrogen atoms. The second deformation removes these dislocations from their atmospheres and causes a greater density of dislocations to be built up, characteristic of the greater degree of work hardening. At the same time, the lines of solute atoms are rendered unstable by being separated from their dislocations, and are also broken up to some extent by the plastic deformation. Accordingly, the atoms in these lines become dispersed and start migrating to the new dislocations, thus causing a second ageing change characteristic of the 42% reduction.

COMPARISON WITH STRAIN AGE-HARDENING

To relate the resistance change to strain age-hardening, the hardnesses of some wires were examined during ageing. The same working and ageing treatments were used as before, and the hardness was measured with a Vickers diamond indenter. Figure 6 shows the hardening curve obtained at 31° C., with the resistance curve included for comparison. Apart from the fact that the hardness change starts more slowly, the two curves are similar, and there can be little doubt that they are both caused by the same structural change in the iron. This is confirmed by tests at other temperatures; the times for half-completion of the hardness change were found to be 900, 250, 100, and 23 min., respectively, at the temperatures of 21°, 31°, 41°, and 51° C. These results give an activation energy of 22,000 cal./mol., which is the same as for the resistance change, to within the limits of experimental error (Fig. 3, curve *C*).

DISCUSSION

The facts that the observed resistance change (*a*) has the same activation energy as for the diffusion of carbon and nitrogen in α-iron, and (*b*) correlates

closely with strain age-hardening, which is already known to be due to carbon and nitrogen,[7, 9, 22] show that the change is concerned with the migration of these elements in the cold-worked metal. Since the resistance decreases on ageing, the atoms evidently migrate from random solution, where they are known to scatter electrons effectively, to some other arrangement where they are less effective. Although this conclusion is consistent with the dislocation theory, the question arises of whether it could be interpreted in some other way. Thus, it might be argued that cold working causes carbon and nitrogen to precipitate from solid solution. However, it must be remembered that the process of segregation to dislocations can itself be regarded as a special kind of precipitation ; in fact, one could regard the dislocation theory as providing a detailed atomic mechanism showing how 'precipitation' can be brought about by cold work. Considered as a specific mechanism of precipitation, and admitting the possibility of alternative mechanisms, the dislocation theory can be tested only by comparing its detailed predictions with experiment. The evidence gained in the present work, particularly the magnitude of the observed resistance change, the indifference of the time of change to the amount of deformation, and the results of the repeated ageing experiment, supports the theory.

If the theory is correct, measurement of the change of resistance on ageing ought to provide a more accurate method of determining the density of dislocations than has hitherto been possible. By working with fine, short wires, so that the early part of the ageing process can be followed, and by using a more sensitive method of measuring the change of resistance, the relation between the density of dislocations and work hardening could be determined accurately. The uncertainty of the factor relating the resistance change to the amount of solute transferred could be removed by experiments with controlled and known traces of solute. With such an apparatus it would be worth while examining single crystals to see whether Bhatia's prediction of the anisotropy of the resistance change (depending on the orientation of the dislocations relative to the direction of current flow) could be verified.

Acknowledgments

This work forms part of a programme supported by the Department of Scientific and Industrial Research through the National Physical Laboratory, to whom grateful acknowledgment is made. The authors also wish to thank Professor D. Hanson for his encouragement and help, Professor N. F. Mott, Mr. B. A. Bilby, and Dr. A. B. Bhatia, for valuable discussions, and Mr. E. Marks and Messrs. Richard Thomas and Baldwins, Ltd., for undertaking the chemical analysis of the wire.

References

1. N. F. Mott and F. R. N. Nabarro : " Report of the Bristol Conference on the Strengh of Solids," p. 1. London, 1948 : The Physical Society.
2. A. H. Cottrell : *Ibid.*, p. 30.
3. G. I. Taylor : *Proceedings of the Royal Society*, 1934, series A, vol. 145, p. 362.
4. J. M. Burgers : *Proceedings of the Royal Academy of Science, Amsterdam*, 1939, vol. 42, p. 293 ; *Proceedings of the Physical Society*, 1940, vol. 52, p. 23.
5. F. R. N. Nabarro : " Report of the Bristol Conference on the Strength of Solids," p. 38. London, 1948 : The Physical Society.
6. A. H. Cottrell and B. A. Bilby : *Proceedings of the Physical Society*, 1949, series A, vol. 64, No. 4, p. 49.
7. C. A. Edwards and others : *Journal of The Iron and Steel Institute*, 1940, No. II, p. 199p.
8. J. L. Snoek : *Physica*, 1941, vol. 8, p. 734.
9. J. R. Low and M. Gensamer : *Transactions of the American Institute of Mining and Metallurgical Engineers*, 1944, vol. 158, p. 207.
10. J. L. Snoek : *Loc. cit.*, p. 711.
11. L. J. Dijkstra : *Philips Research Reports*, 1947, vol. 2, p. 357.
12. W. Köster : *Archiv für das Eisenhüttenwesen*, 1928–29, vol. 2, p. 503.
13. F. Seitz and T. A. Read : *Journal of Applied Physics*, 1941, vol. 12, pp. 100, 170, 470, and 538.
14. W. F. Brown : *Physical Review*, 1941, vol. 60, p. 139.
15. J. S. Koehler : *Physical Review*, 1941, vol. 60, p. 397.
16. C. F. D. Yensen : *Journal of the American Institute of Electrical Engineers*, 1924, vol. 43, p. 455.
17. E. S. Davenport and E. C. Bain : *Transactions of the American Society for Metals*, 1935, vol. 23, p. 1047.
18. T.-S. Kê : *Metals Technology*, 1948, Technical Publication No. 2370.
19. D. Polder : *Philips Research Reports*, 1945, vol. 1, p. 5.
20. A. B. Bhatia : *Proceedings of the Physical Society*, 1949, series B, vol. 62, No. 4, p. 229.
21. W. L. Bragg : *Transactions of the North-East Coast Institution of Engineers and Shipbuilders*, 1945, vol. 62, p. 25.
22. L. B. Pfeil : *Journal of The Iron and Steel Institute*, 1928, No. II, p. 167.

Surface Film Stripping of Metals

Work on film stripping now being developed at the Chemical Research Laboratory, of the Department of Scientific and Industrial Research, includes removal of the invisible protective film on polished stainless steel.

Film stripping by chemical methods was originally applied to iron, and made use of an aqueous solution of iodine which attacked the iron just below the surface of the overlying oxide, permitting the film to be loosened and stripped. The present work makes use of a solution of iodine in anhydrous methyl alcohol, in an atmosphere of nitrogen, which eliminates contamination of the film by water and oxygen. Reinforcement of the film with a coating of synthetic resin enables it to be removed in one piece, and mounted on glass for examination.

The composition of the films may be conveniently studied by electron-diffraction, both of the oxide film *in situ* and of the stripped film. This method has already been used to study the thickness and composition of oxide films formed on iron and steel at various temperatures, and of the film formed on polished stainless steel.

Possible applications with ferrous metals, are the study of the oxide films produced by different methods of physical and chemical preparation. The study of metal films has a direct bearing on finishing, for chemical pretreatment of steel is used to increase adhesion of the finishing coat. The technique has also been used to isolate films produced by the action of corrosion inhibitors.

DISCUSSION ON PAPERS

DISCUSSION AT MEETINGS · WRITTEN CONTRIBUTIONS

AUTUMN GENERAL MEETING, 1948

The discussion on the first four papers presented at the meeting (" Cowper Stoves " ; " Magnetic Concentration Experiments " ; " Statistics in Heavy Industry," and " Oxygen Enrichment in the Side-Blown Converter ") was published in the June, 1949, issue of the *Journal*.

The discussion on three further papers follows, and that for the remaining papers will be given in the August issue of the *Journal*.

Discussion on the Paper—

DEVELOPMENT OF AN IMPROVED BASIC BESSEMER STEEL*

by H. A. Dickie

Dr. **H. A. Dickie** (Stewarts and Lloyds Ltd., Corby) presented his paper.

Mr. **A. J. K. Honeyman** (Colvilles Ltd.) : The author has stolen some of my thunder, in presenting the paper, by saying that the best of the Bessemer steel approaches the worst of the open hearth ; that is just the point that I was going to make. On p. 364 he says " As a result of the tests, it has been fully established . . . that when the phosphorus and nitrogen contents of Bessemer steel are reduced to values similar to those of the best quality open-hearth steel the differences between the steels disappear." Exactly ; but the author's own figures show that the nitrogen content of his improved Bessemer steel does not approach that of the best open-hearth steel. The figures given on p. 370 show that the average nitrogen is 0·0085%, and the figure for what he says represents the best practice is 0·0073% ; but the solubility of nitrogen in steel at 0° C. is 0·0075%,[1] and that is the absolute maximum, not the average, that you can have in steel to get rid of the effects of age hardening due to nitrogen. The samples which the author has quoted to show that his improved Bessemer steel approaches open-hearth steel merely show that he has got almost half way. I do not wish, however, to minimize the improvement which he has made in basic Bessemer steel by this process. With reference to the results quoted in Tables IV and V, phosphorus is generally considered to improve the corrosion resistance of mild steel. It may be, therefore, that the results quoted are due in large measure to differences in carbon content. Would the author state the complete analyses of these steels ?

Dealing with embrittlement, on p. 362 the author says, " Before the ultimate causes of embrittlement can be isolated and defined, it is necessary to devise, or adapt, testing methods capable of reproducing the effects, and which allow an evaluation of the factors involved to be made." He then makes some criticism of the Izod impact test, and describes a cupping test which certainly does not do what he says is necessary. The cupping test is merely an acceptance test, and simply says whether the material comes up to a certain standard or not ; it cannot distinguish between two steels which both pass the test and say whether one is better than the other. The Izod test, on the other hand, is designed to distinguish between different steels and to place them in order of merit ; it is a very much more informative test than the cupping test or the other test devised by the author.

His discussion of embrittlement refers entirely to embrittlement accompanied by marked hardening, but one of the great problems of mild steel, today, is that form of embrittlement not shown by an increase in hardness or by any of the usual mechanical tests. It is this form of embrittlement which has restricted the use of Bessemer steel ever since open-hearth steel began to compete with it. Has the author done any work on these steels recarburized, say, to 30 tons/sq. in., and if so, how did their impact properties compare with those of open-hearth steel of similar tensile strength ?

Dr. **P. M. Macnair** (Richard Thomas and Baldwins, Ltd.) : I agreed with Dr. Dickie's opening statement that " The differences in the intrinsic properties " of Bessemer and open-hearth steel " are not all weighted in favour of either."

Efforts have been made to lower the nitrogen content of basic Bessemer steel at Ebbw Vale. In the initial stages iron ore was used in the same way as at Corby. It was found to be effective, but now a scrap steel addition is preferred, shortly after the carbon blow develops its full force. The action is similar, reduction of the final temperature of the metal, and a reduction of nitrogen from around 0·015% to an average of 0·008%.

The author has pointed out that basic Bessemer steel can be finished as an open-rimmer, a controlled rimmer, or as a killed steel. In developing the controlled rimmer with a plug top, the author has produced an ingot which is an excellent compromise between the advantages and disadvantages of the more usual open-rimmer and killed steel. It has the good surface of the rimmer, but the internal segregation is considerably reduced.

This raises one thought in my mind, on which I would like to ask the author's opinion. In order to get a

* *Journal of The Iron and Steel Institute*, 1948, vol. 159, Aug., pp. 360–375.

controlled rimmer, I feel it is desirable to make a steel which, in the ordinary open mould, would be a box-hat or a sinking-rimmer type. If my suggestion is correct, it would seem to be just one more advantage that Bessemer steel has over open-hearth steel.

The author has been careful to present his facts and deductions in a straightforward manner, and has made us realize that basic Bessemer steel can be made regularly, with satisfactory properties.

Dr. **A. H. Leckie** (British Iron and Steel Research Association): I was very glad to hear Dr. Macnair mention the scrap process used at Ebbw Vale, because as far as I know the only two processes worked in this country, today, to get low-nitrogen basic Bessemer steel, are the process which the author has described and the control of temperature by the use of scrap, as used at Ebbw Vale. It is not clear from the paper whether the author considers that the effect of ore in giving lower nitrogen is due to getting oxygen free from nitrogen, or due entirely to the cooling effect. If it is purely a temperature effect, one would think that the process used at Ebbw Vale would have a slight advantage, since the temperature is controlled by a material which gives a greater iron yield. It would be very interesting to know whether the effect is due entirely to temperature, entirely to freedom from nitrogen in the oxygen supply, or to both.

Dr. Newell, in a previous discussion, pointed out that oxygen enrichment, in the proportions which can be used in practice, is not going to eliminate nitrogen to any great extent. People talk too glibly about putting oxygen in the converter and getting rid of nitrogen troubles; as far as is known, the highest degree of oxygen enrichment it is practicable to apply in a bottom-blown vessel still leaves a blast containing a lot of nitrogen. Oxygen enrichment may help in another way, because it gives increased temperature and allows the use of even more ore. Does the author contemplate using oxygen together with ore in this process?

I should also like to ask whether he has to take any special precautions in drying the ore and scale used, and, if not, are any troubles with hydrogen experienced?

Professor **W. J. Geller** (Aachen University): I can recall, only too well, the trouble we have had in Germany, and are having still, in developing an improved basic Bessemer steel with a low content of nitrogen. The figures of the nitrogen content given in the paper may be compared with the best results obtained in Germany for the production of an improved steel. It has, in most cases, been done in a similar way* by adding iron ores, using a converter with a worn lining, and taking only the lower part of the ingots when it was necessary to meet more stringent requirements.

The greatest snag in the process, in my view, is the skull in the ladle. The temperature of the steel, having a decisive influence on the pick-up of nitrogen, must be kept as low as possible, so that, exaggerating the facts, one might say that the lowering of the nitrogen content was proportional to the weight of the skull. The low temperature needed is a special handicap in bottom pouring. Does the author have the same trouble, and if so to what extent? I should be interested to know some details about the number, arrangement, and size of the tuyeres. The author has shown that the nitrogen content of the steel increases with the age of the bottom. I would say that this, apparently, does not apply to all converters. Eichholz and Kootz, in their extensive study of the basic Bessemer process, could find no influence of the age of the bottom. Böddeker[2] found a slight trend in the direction which the author has mentioned, but Wilms and I[3] have found indications of the reverse.

With regard to the influence of the nitrogen content in the iron, I should like to refer to the statement of Kootz,[4] as well as that of Daeves, Frerich and Holtmann,[5] that the nitrogen content of the steel increases with that of the iron. In the presence of vanadium, chromium, and silicon, one must be especially careful in selecting a suitable method of analysis to determine all the nitrogen in the iron.

The figure given for the cooling effect of the iron ore as compared to scrap agrees roughly with our own experience, but on the basis of a theoretical calculation, the cooling effect of the ore should be 3 to 4 times as high as that of the scrap. I wonder whether a satisfactory explanation could be given for this discrepancy. One explanation might be that in the presence of iron oxide more of the carbon is burnt to carbon dioxide, but I have no proof of that.

Other processes have been developed for the production of an improved basic Bessemer steel in Germany. In one of our steel plants it was done by increasing the pressure of the blast to about 1½ times the normal pressure (up to 2·8 atm.) which resulted in a shortened blow.[6] Cooling was achieved by the addition of scrap; but it is claimed that the temperature may be kept higher than in the other process although the same nitrogen content is obtained.

Another process for which the nitrogen figures may be compared with those of a good open-hearth steel, and which leads to a high-temperature steel, has been worked out by the use of a side-blown converter.[7] The process is especially useful in bottom-pouring, though it has the bottleneck of a prolonged blow—up to ¾ hr. for a 30-ton charge.

The last development started from experiments with a drum-shaped converter. This led to an elliptical variation of the normal converter and a special arrangement of the tuyeres in the bottom. For further details I would refer you to the recent paper by Kootz and Gille.[8]

It would be extremely valuable if Mr. Harrison, Dr. Newell, and Mr. Hartley could give some details about the effect of oxygen enrichment, with the side-blown converter, in influencing the nitrogen content of the steel. The side-blown converter process is known to be especially liable to lower the nitrogen content[7, 9, 10], and as Wentrup and Reif[9] have shown, it is possible to lower the nitrogen content well below that of the iron even with normal air and exceedingly high temperatures. In one of their experiments with an acid converter they had only traces of nitrogen, with a temperature, if I remember rightly, of about 1800° C.

Dr. Ir. **H. G. Geerlings** (Royal Dutch-Shell Group, Amsterdam): At the Amsterdam Laboratory of the Royal Dutch-Shell Group we received some samples of this steel and carried out a few experiments on it in comparison with open-hearth steel, which we generally use as line pipe. We were especially interested in the ageing susceptibility of the steel, as the normal tensile, flattening, hardness tests, and so on, do not reveal sufficiently clear differences.

To determine the ageing susceptibility of LNP steel, a sample of it was stretched 10% and heated for 1 hr. at 200° C. The Charpy impact value was determined and compared with that of the untreated sample. As tests at room temperature give insufficient information, the impact value was determined as a function of the temperature. From the area of scatter, only the beginning is indicated

* The process has been developed by the Vereinigte Stahlwerke, A.G., in Hamborn and Hoerde, and because of the low content of phosphorus and nitrogen the steel has been called LPN steel.

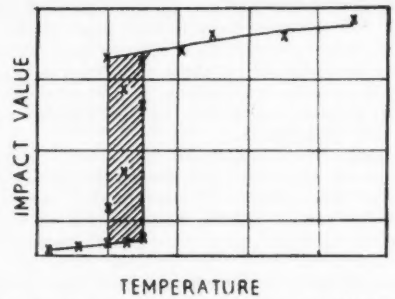


Fig. A—Temperature/impact-value diagram

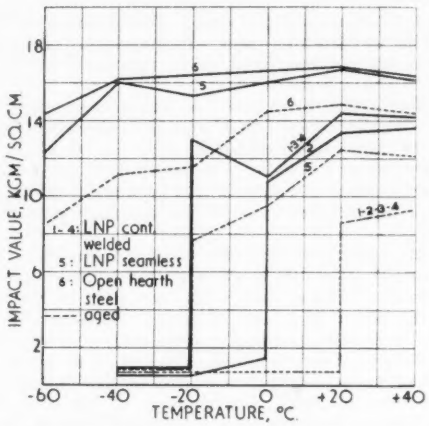


Fig. B—Susceptibility for ageing of LNP and O.H. steel

in the temperature/impact-value diagram (Fig. A). The complete results of these tests, in which the dimensions of the test bars were : 10 × 4·2 × 55 mm. ; notch depth, 3 mm. ; radius, 1 mm., are given in Fig. B. Samples 1, 2, 3, and 4 are of LNP steel pipe, continuously welded ; No. 5 is of LNP steel, seamless, and No. 6 is of open-hearth steel. It will be seen that with the LNP steel there is a drop in the impact value after ageing at — 20° and + 20° C. There is a difference between the LNP steel from the continuously welded pipe and from the seamless pipe. Sample No. 5 shows no drop of impact value above — 60° C., in the condition as delivered, but after ageing there is a drop at — 20° C. The difference must be attributed to the difference in grain size.

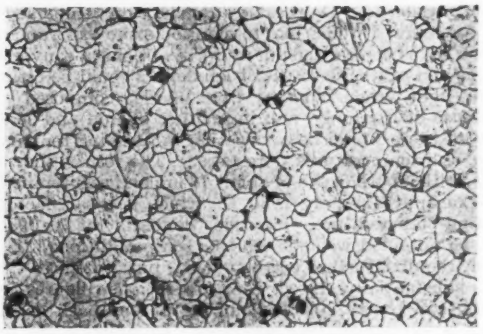


Fig. C—Structure from seamless pipe. Etched with nital × 150

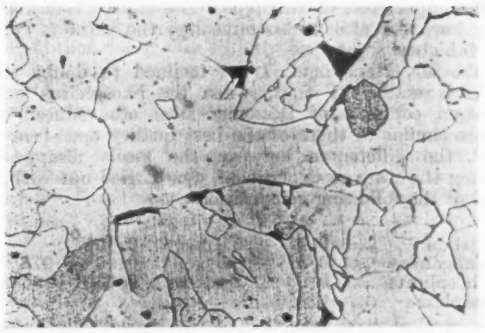
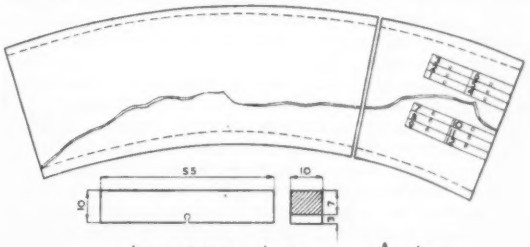


Fig. D—Coarse structure from continuous-welded pipe. Etched with nital × 150

Fig. E—Ageing phenomenon in cold bent pipe

Figures C and D show the different structures : Fig. C the seamless pipe ; Fig. D the very coarse structure from the continuous-welded pipe. Grain-size reduction results in a shifting of the transition area to lower temperatures. It will be seen that the results obtained with open-hearth steel, both in the condition as delivered, and after ageing, are better than those obtained with LNP steel, which is in agreement with the author's observations. As Figs. 32 and 33 of the paper show continuous curves, I should be glad to know how many times the author's experiments were repeated, and whether an area of scatter was not noticed in the Izod impact tests.

That the occurrence of ageing phenomena is not imaginary in the case of cold bent pipes of converter steel is clear from Fig. E. This pipe was installed during the war and is of bad quality converter steel. At the time when there was a shortage of open-hearth steel, this pipe was built into one of our refineries, where it gave rise to brittle fracture. Figure F shows the result



Impact test values				Analysis
As received		Stress annealed		
1. 3·5 kgm/sq.cm.		4. 4·4 kgm/sq.cm.		C - 0·13%
2. 1·7 "	"	5. 8·9 "	"	Mn - 0·41%
3. 4·0 "	"	6. 11·4 "	"	Si - 0·19%
7. 2·0 "	"	10. 10·3 "	"	P - 0·083%
8. 2·4 "	"	11. 10·9 "	"	S - 0·018%
9. 1·8 "	"	12. 11·8 "	"	N - 0·015%

Fig. F—Results of impact test on cold bent pipe shown in Fig. E

of the impact test on this pipe. The impact values are very low, and after stress annealing the value is very much higher.

Like Mr. Honeyman, I am inclined to doubt the author's statement that, when the phosphorus and nitrogen contents of Bessemer steel are reduced to values similar to those of the best quality open-hearth steel, the differences between the steels disappear. During the war, an experiment was carried out with a steel of the following composition :

	%		%
Carbon ...	0·07	Nitrogen	0·004
Manganese ...	0·37	Oxygen	0·008–0·015
Phosphorus ...	0·03	Hydrogen	0·0004–0·0009
Sulphur ...	0·016		

That steel showed reduction of the impact value, after ageing, from 17 to 1 kgm./sq. cm. Undoubtedly nitrogen has a great influence, but there are a number of additional factors which considerably affect the ageing susceptibility of the steel, *e.g.*, grain size.

Mr. **A. Robinson** (Appleby-Frodingham Steel Co.) : I have recently been examining the Thomas process in this country and elsewhere, and I have been surprised at the improvement in Bessemer steel in recent years. From what I can see, it is going to be considerably improved in the future.

The additive effects of phosphorus and nitrogen are interesting, and the research opens up the prospect of using this steel for purposes for which it has not been regarded favourably in the past. The comparisons made are not, in my view, quite fair to open-hearth steel, as the comparative samples were somewhat higher in carbon ; but it is well known that it is much easier to make a low-carbon steel in the Thomas process than in the open-hearth.

It is interesting to note the convergence of the two processes : the open-hearth is now experimenting with air and oxygen to get the oxidation, and the Bessemer process with oxides to perform the oxidation. Shall we eventually come to a cross-bred furnace ?

The absorption of nitrogen depends on the time, pressure, and temperature, and it is difficult to get them all working in the same direction. The process described shows a variation with the age and wear of the vessel. The use of smaller charges or a larger vessel would have the same effect on the output obtainable from the plant, but a blowing time of 50% longer would not make a one-third difference in the output, and because account has to be taken of the time spent on tapping, fettling, charging, and so on, it would probably amount to one-sixth, and, as most Bessemer shops have an enormous potential output which is seldom used fully, this would not be so serious a matter.

With shallow blowing the nitrogen can be reduced from, say, 15 or 18 points down to 10 or 12 (that is the experience that I have had recently), with an increase in, or no loss of, temperature, due to the combustion of part of the carbon monoxide in the converter, and that is confirmed by side-blown converter practice. Making use of oxygen, this effect may be increased, with a time shortening, due to the use of oxygen allowing the addition of oxides to keep the charge cool and preventing or lowering nitrogen pick-up, and all these small amounts may add up. The pressure, and the temperature, and the time, can then be made to work in the same direction. There seems no reason why the same nitrogen as in open-hearth steel should not be produced regularly, and a very soft steel produced with the added advantage of a lower carbon and freedom from unwanted metals derived from the scrap.

A modification of the cylindrical type of converter used in copper refining may be the answer. The prospect of making a very low carbon, phosphorus, and sulphur steel with low nitrogen from British ores should be of interest to many people. This type of steel should be thoroughly investigated by those concerned in pressing, as it is much easier to produce a first-class rimming ingot with a thick sound skin by the Thomas process than by the open-hearth.

Mr. **J. Glen** (Stewarts and Lloyds, Corby) : The steelmaker is faced with what are frequently two opposing problems in the question of steel quality *versus* such things as production, yields, consumption of material and so on, and I expect that the steelmakers present are asking themselves how the process described does affect these things. On an industrial scale there must be a compromise between these important factors. We cannot leave either side weighted to a ridiculous extent ; in other words, we cannot have yield, for instance, at any cost in quality, nor, alternatively, ultra-exceptional steel quality at any cost in yield. If, however, the quality improvement is substantial, as in this case, we should expect to pay a little for it.

The first thing to say is that the process definitely works. We can make a steel such as is described in the paper, of very low phosphorus, very low sulphur, and low nitrogen, of the order of 0·006–0·007%, and we can make such steels at will, and with low incidental impurities, and make them by a process naturally adapted to make the best type of rimming ingot

On the question of output or production, as fitted into the existing operation, there is no deleterious effect on production. Where one scale addition is made, it takes the place of a normal scrap addition, and the blowing time is about 2 min. shorter than with the ordinary blowing. Where we make the two scale additions, the reduction in blowing time is offset by the time required to make the additions, but there is no effect on bulk production. Although the theoretical considerations given in the paper seem weighty and complicated, the application of the process is extremely simple. There is no interruption to crane availability, which is important in Bessemer work, nor any alteration to the routine performed by the steelmaking personnel. There seems to be a slightly greater tendency to nose skulling, but this is not serious.

With reference to Dr. Macnair's statement regarding the production of low-nitrogen steel by control of temperature with scrap addition, we have tried this too, but, at equal finishing temperatures, we find that the scale-blown steel is at least 0·002% less than that made with scrap additions. This is considerable on values of 0·01%, or under. The blast in Dr. Macnair's case is less subject to deoxidation because the blow starts with 1·5% phosphorus further diluted by heavy scrap additions.

With regard to Dr. Leckie's remarks on temperature and pressure, I would like to mention some experiments which have been carried out at Corby. Using a converter at the end of its life so that bath depths could be estimated, light-weight charges were made, composed in one case of iron only, and in another of iron and scrap so that there would be a variation in finishing temperature, and temperatures were taken. The results showed that, with a bath depth of about 18 in., and at a temperature of approximately 1630° C., which is a satisfactory casting temperature, nitrogen contents of 0·008% can be obtained without scale or scrap additions. This work requires further corroboration, but I believe that it is substantially correct that low nitrogen figures can be obtained with straight blowing at low bath depths with good casting temperatures. In practice we feel that temperature is a very important factor when the other factors are wrong, but where the other factors are

correct it does not have a substantial effect and this is in accordance with Dr. Dickie's paper. We think that the pressure effect is the important factor, bearing in mind that the partial pressure of nitrogen can be increased by bath depth, blast deoxidation, or even by temperature.

The liquid-steel yield from the modified process for the year, to date, is about 1·5% less than that from the ordinary process. When considering this statement, certain things, including the definition of yield, should be taken into account. Firstly, the difference between 0·05% and 0·03% phosphorus in the ordinary process means about 2% lower yield ; that is, if we blow to 0·03% phosphorus in the ordinary process, then the yield from both the modified and the ordinary processes would be about the same. Also, in calculating the yield on the modified process, the scale addition has been included in the materials input side at its iron value. This is roughly equivalent to 3% of the charge. If we omit the scale addition altogether in our yield calculations, then the liquid-steel yield, on this year's performance, would be roughly 1·5% higher than with the ordinary Bessemer process. The purpose of the ore or scale addition is to supply oxygen rather than iron, and if some other oxide had been used it would probably have been treated as, say, lime or other flux, and ignored for yield purposes.

The ingot yield for the modified process for the year, to date, is about 2·5% lower than in the normal process. There has been a tendency to cold casting, which affects ingot yield by ladle skulling, but this is not inevitable. A low nitrogen content can be obtained with casting temperatures which avoid skulling. There seems to be, in practice, a critical temperature well above that at which skulling would occur, at which really low nitrogen of the order of 0·005–0·006% can be obtained. Any further reduction in casting temperature, as might be expected because of the low nitrogen figures quoted, gives no further improvement in nitrogen content. There has been a gradual evolution on the practical side of the process and we now prefer a rather higher silicon than was formerly considered ideal for ordinary blowing. Under these conditions we can obtain both a low nitrogen content, and a correct casting temperature which does not affect the yield.

The cropping yields in the blooming mills for the rimmed ingots made by the modified process are about 2% better than for those made by the ordinary process.

In view of the general remarks regarding increased temperatures to be obtained by the use of oxygen, I should like to draw attention to the fact that, in the Bessemer process, any big increase in temperature would only be required for carrying scrap, and at present we often use it in the reverse way, using scrap for cooling down. We do not want excessively high casting temperatures on the practical side because of the adverse effect on steel, moulds, ladles, etc.

With regard to the effect of the process on lining life and refractories, it is difficult to make any definite pronouncement, since we have never operated a converter 100% on scale-blown steel, but we have not noticed any deterioration in lining life, when making 30% of the output as scale-blown steel over a long period. We have an opinion, it is only an opinion, that the process actually improves the lining life. It can be shown that the MgO in the slag rises in the after-blow, in which period the slag volume and the temperature are also increasing. This indicates that the converter lining is more affected in that period, and possibly shows a temperature effect.

It can also be shown that the MgO in the slag increases with increasing iron content of the slag, and undoubtedly the iron is higher in the slag from the modified process than in that from the ordinary process ; but we think that, where a high iron occurs in the ordinary process, it is probably only indicative of high temperature, though not necessarily so. My personal opinion is that, so far as the Bessemer process is concerned, the effect on the lining is largely a solution loss caused by high temperature, and, if this is so, it would be expected that the new process, by reducing the temperature of the metal in this period of the blow, would give improved lining life.

While the frequency curves in Fig. 19 were true for the time when the author made them, and represent what were normal phosphorus and sulphur results two or three years ago, it should be noted that for the last eighteen months or two years, owing in no small measure to the experimentation which was carried out under the guidance of the author, the frequency curves for sulphur in particular, in ordinary Bessemer steel, should more or less coincide with the curve for modified steel. The general run of phosphorus is also lower than is indicated by that curve. The whole production at Corby is made to a maximum of 0·040% sulphur, and much of it to considerably lower specifications.

With reference to the author's remarks on embrittlement without notches, is this in any way connected with locked-up stresses as a result of the deformation to which the steel has been subjected ?

Mr. **M. W. Thring** (British Iron and Steel Research Association) : This paper raises a number of interesting points which one would like to take up in a more fundamental way than is possible in a works. The author refers to the special cupping test that he found necessary to test the property which comes from the presence of nitrogen in the steel, and Mr. Honeyman said that he preferred the Izod test. The fundamentalists at the universities regard even the Izod test only as one which divides things into brittle and not brittle, and not as having a quantitative value.

One would like to see the physics of this question of brittleness carried a good deal further. For instance, Cottrell at Birmingham has produced a theory about the presence of carbon in steel preventing dislocations from wandering and so producing the sharp yield-point of steel, and in his theory nitrogen plays exactly the same rôle as carbon. So it is just possible that this criterion of work-hardening, $P + 5N$, might be extended, from the point of view of mechanical properties, to $P + 5N + xC$, so including carbon in the same formula. Does the author think that nitrogen and carbon have the same effect, or are they quite different from this point of view ? According to Cottrell's theory, they both have the property of anchoring the dislocations and so preventing plastic flow ; ageing is then considered to be due to the diffusion of carbon atoms through the metal to re-anchor the dislocations. Dr. Dickie's results could be very profitably married to Cottrell's theory, and there is a good deal of interesting fundamental work to be done in that direction.

The second fundamental point is the physics of gas–metal contact in the bottom-blown and side-blown Bessemer. One view is that it is simply a question of time of contact and area of contact, in which case his small-scale experiment gives a formula which could actually be applied numerically to the full-scale plant, both side-blown and bottom-blown, if one knew the area of contact and the time of contact.

An impression I have from looking into a bottom-blown Bessemer converter at Corby, is that whole chunks of metal are being thrown about inside, and that there is a tremendous area of contact and that this is a reason why the side-blown, with a relatively small area of contact, gives so much less nitrogen pick-up.

This does not explain, however, how there can be sufficient contact to get the carbon out without getting the nitrogen in.

It would also be very interesting to test, by means of a model, whether it is the top, bottom, or mean diameter that affects the contact in the shallow-bath Bessemer shown in Fig. 17.

Finally, Dr. Dickie's fascinating paper led him to speculate on the factors which, in the period when the steel industries of the world were expanding so rapidly, decided that the open-hearth furnace outstripped the Bessemer. The physical properties of the steel must have been important, but probably other economic factors, like the relative amounts of scrap and pig used, or the amount of metal lost, were also very important.

CORRESPONDENCE

Monsieur **Y. Dardel** (Centre de Documentation Sidérurgique) wrote: The author has presented a significant contribution to the manufacture of an improved basic Bessemer steel and I have been very glad to have the opportunity to read his results.

It seems to me, however, that the analysis of nitrogen absorption is not entirely satisfactory.

The solubility of a bi-atomic gas in a liquid metal (*see* Smithells,[11] and Fowler and Smithells[12]) does not vary proportionally to T^5 but generally to $e^{\frac{A}{T}}$ The experiments of Chipman and Murphy,[13] and chiefly those of Kootz,[14] confirm this point for the nitrogen–iron system (*see* Fig. G).

Besides, according to these authors,[13, 14] the absorption speed is proportional to the difference existing between the actual amount of dissolved nitrogen [N] and the amount at the saturation $[N]_m$. One has (*see* Fig H):

$$\frac{d[N]}{dt} = K_1([N]_m - [N]) \quad \ldots\ldots\ldots\ldots (1)$$

or

$$[N] = [N]_m\,(1 - e^{-k_1 t}) \quad \ldots\ldots\ldots\ldots (2)$$

At the beginning of each run of Chipman's experiments, the absorption rate was somewhat irregular, because of the adjustments in power that had to be made in order to bring the charge to the desired temperature.

The absorption, therefore, runs like a monomolecular reaction which can be easily explained, while the hypothesis, according to which the absorption would be proportional to the square root of the time, remains without clear explanation.

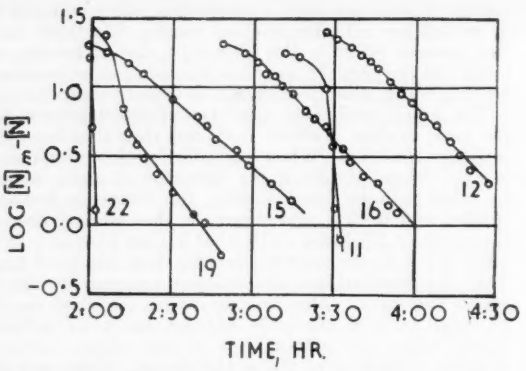


Fig. H—Logarithmic plot of rate of solution. According to Chipman and Murphy[13]

Taking into account the law of solubility of a gas in a liquid metal, as well as the solubility values already published, it is possible[15, 16] to estimate the solubility of nitrogen in iron, and to discuss the variation of nitrogen amount during blowing.

According to Chipman and Murphy[13] as well as Kootz[14] the nitrogen solubility m obeys Sievers' law:

$$m = K_s\sqrt{p_{N_2}} \quad \ldots\ldots\ldots\ldots (3)$$

where K_s is a constant depending only upon temperature, and p_{N_2} the partial pressure of nitrogen in the gas phase.

According to Fig. G, the nitrogen solubility varies very little with temperature and can be represented by the following equation, in relationship with temperature and the partial pressure of nitrogen:

$$\log m = -\frac{1{\cdot}161}{T} - 0{\cdot}721 + \tfrac{1}{2}\log p_{N_2} \ldots\ldots (4)$$

m being given in weight per cent., and p_{N_2} in atmospheres.

Kootz has also shown that the addition of phosphorus or carbon has the same effect on the nitrogen solubility (*see* Fig. J). At a temperature between 1600 and 1630° C. the diminution of nitrogen solubility in relation to the amount of Fe_3C or Fe_3P would be equal to:

$$\log m = 0{\cdot}0092 \times Fe_3C \text{ (or } Fe_3P)\,\% - 1{\cdot}330 \ldots (5)$$

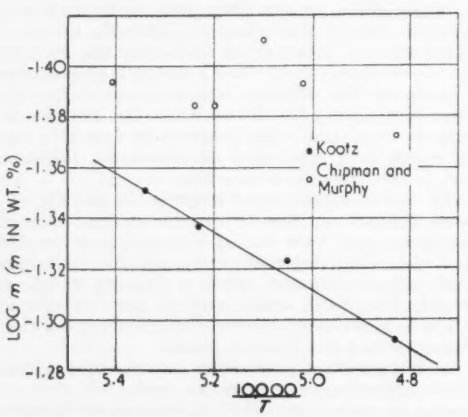


Fig. G—Solubility of nitrogen in pure iron. These are taken from results obtained by Chipman and Murphy,[13] and Kootz[14]

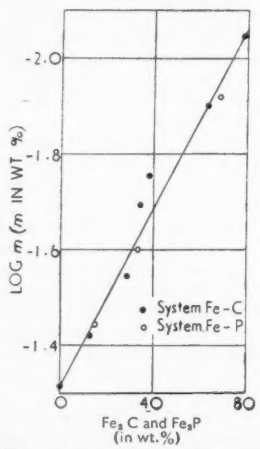


Fig. J—Solubility of nitrogen in the systems Fe–Fe_3C; Fe–Fe_3P at 1600–1630° C. According to Kootz[14]

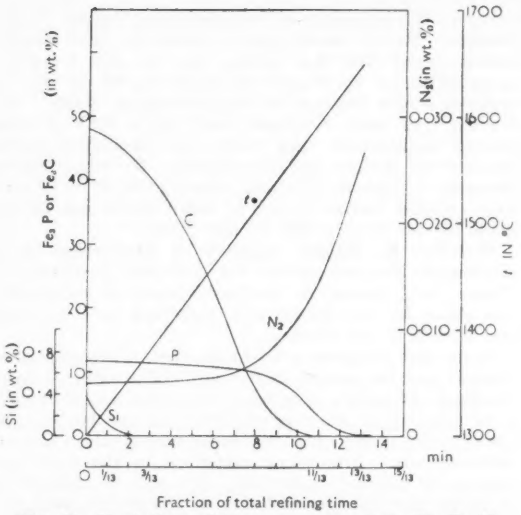


Fig. K—Variation of the amounts of C, Si, P, N_2 during refining

The effect of temperature on the nitrogen solubility in Fe–C and Fe–P systems has not been studied. However, by means of the equations (4) and (5), and the general knowledge of the laws of gas solubility in metals, the solubility of nitrogen can be roughly estimated in a much wider range of temperature, as given by equation :

$$\log m = -\frac{1{\cdot}161}{T} - 0{\cdot}721 + \tfrac{1}{2}\log p_{N_2} - 0{\cdot}0092\,(Fe_3C + Fe_2P) \quad \ldots\ldots\ldots (6)$$

The experiments of Chipman and Vaughan,[17] as well as those of Karnaoukhov and Morozov,[18] show that, in first approximation, the effect of silicon on the nitrogen solubility can be neglected when the silicon amount is inferior to 1%.

Therefore, the equation (6), although only roughly representing the conditions of the equilibrium of the system Fe–C–N_2, has the advantage of being valid for pig iron as well as for steel, and allowing us to discuss, at least semi-quantitatively, the absorption of nitrogen in a basic converter.

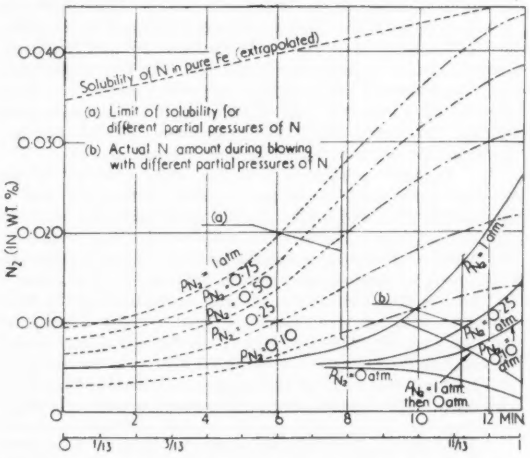

Solubility of N in pure Fe (extrapolated)
(a) Limit of solubility for different partial pressures of N
(b) Actual N amount during blowing with different partial pressures of N
N₂ (IN WT %)
12 MIN

Fig. L—Variation of nitrogen amount and solubility during refining

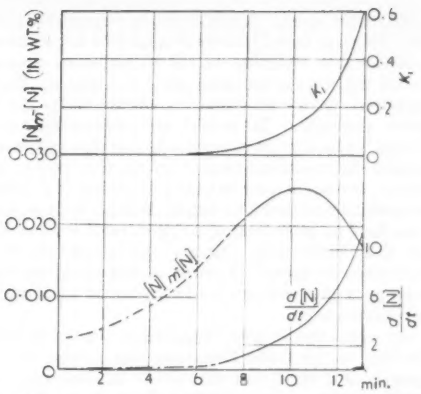


Fig. M—Variation of $[N]_m - [N]$, $d[N]/dt$ and K_1 during refining

In Fig. K the average values of the amounts of C, Si, N_2, and that of temperature are outlined, while the resulting curves of nitrogen solubility are shown in Fig. L. In order to make the effect of temperature distinct from that of the composition, the solubility curve of nitrogen in liquid iron has been extrapolated down to 1300° C. According to Fig. L, the effect of composition is much greater than is usually admitted.

According to the author the nitrogen solubility in pig iron would be 0·008 to 0·010%, in perfect agreement with the value given by the equation (0·009%).

Figure L shows that the increase in the amount of nitrogen dissolved during fore-blowing (curve corresponding to a deep bath) is much smaller than the increase of solubility, and that nitrogen absorption, therefore, is limited by the absorption speed of this gas (equation (1)).

The values of $d[N]/dt$ as well as that of $[N]_m - [N]$ are shown in Fig. M.

The variation of K_1 can be explained by means of the experiments of Naeser[19] and those of Seth.[20] According to Naeser, bubbling of a mixture of $N_2 + O_2$ or $N_2 + CO_2$ through liquid iron does not nitride the metal, while a mixture of $N_2 + H_2$ or $N_2 + CO$ does. In the first case, the oxide film formed at the interface bubble–metal hindered the nitrogen diffusion. The experiments of Seth show that the exit gas of a converter always contains either O_2 or CO_2 during the fore-blow, while it contains almost pure nitrogen during the after-blow.

To summarize, the low nitrogen content of pig iron is due to the amount of carbon and phosphorus, the amount of dissolved nitrogen being not very different from that which corresponds to saturation. As blowing starts, the amount of dissolved gas tends to come in equilibrium with partial pressure of nitrogen in the gas phase ; but, owing to the presence of an oxide film surrounding the bubble during the fore-blow, the reaction is very sluggish. When the temperature has risen enough to dissolve quickly the oxide formed, nitrogen comes directly in contact with the metal and can be absorbed easily.

Regarding the manufacture of low-nitrogen steel. I agree with the conclusions of the author. Other means of obtaining low-nitrogen steel, however, consist of blowing with oxygen-enriched air,[21] or to fore-blow with air and to after-blow with pure oxygen.[22]

Monsieur **R. Perrin** (Société d'Électro-Chimie, d'Électro-Métallurgie et des Aciéries Électriques d'Ugine, France) wrote : I have read, with interest, the paper on the methodical researches carred out at Corby and the good results obtained in improving the quality of the

basic converter steel. A particularly remarkable result, and one which, in fact, Dr. Dickie considers as paradoxical, is the noticeable lowering of the phosphorus content of the metal with an even iron-oxide content of the slag as compared with the contents obtained in the basic converter practice. It would be interesting to know Dr. Dickie's opinion on the reasons for this result which, failing any information from him on this point, is not, I presume, the consequence of the addition of a decidedly higher quantity of lime. It seems difficult to imagine how just the fact of introducing oxygen in the form of iron ore or mill scale may change the equilibrium ratios between slag and metal. A study of the complete analysis of the slags might perhaps lead to a better understanding of the phenomenon.

In any case, the results obtained at Corby, in bringing the quality of the basic Bessemer steel closer to that of the open-hearth steel, are extremely interesting.

Together with our associates, Forges et Aciéries du Nord et de l'Est, we have been working in a different direction : manufacture of killed steels by the Perrin Process, that is to say, by mixing the basic converter steel with a molten slag having a high alumina and lime content, while aiming at a high degree of purity as far as phosphorus, oxygen, and sulphur are concerned, for the production, among others, of deep-drawing steels.

Our main object has been to obtain :

Low-carbon steels of great cleanness
Steels showing great micrographic and macrographic homogeneity in the whole ingot product
Steels with very low ageing
Steels in which the difference between length and transverse properties should be very low as a consequence of very low oxygen and sulphur contents.

The following analyses have been obtained on important tonnages :

C, %	Mn, %	P, %	S, %
0·03–0·08	0·015–0·30	0·015–0·025	0·010–0·017

These analyses have rendered possible the utilization of these grades of steels in France, in competition with Swedish iron.

As regards drawing properties, it has been possible, in certain deep-drawing operations, to reduce the number of passes.

It is true that such results, which go further than those pointed out by Dr. Dickie, have been obtained with killed steels. The difficulties mentioned by Dr. Dickie, weldability and surface defects, have been overcome only by an accurate control of residual aluminium content of the metal, which is made possible by its deoxidation.

Dr. **T. P. Colclough,** C.B.E. (Iron and Steel Control) wrote : The practical operations described in the paper show the development of a technique far more precise and exact than was found possible during the war, on the Continent, where attempts were made to replace open-hearth steel by a Thomas steel, in which the nitrogen content was reduced. These continental attempts achieved only partial success, and the results obtained show very wide variations as contrasted with the positive control attained in the practice described by the author.

It would be of considerable interest if the results obtained in a parallel line of development with the same object, carried out in other works in this country, were made available.

Apart from the data contained in the paper, this report is of importance as marking another step in the removal of misconceptions and ancient prejudices with regard to the basic Bessemer process. In the past, this method of steelmaking deservedly fell into disrepute owing to the irregular and unsatisfactory nature of the product. Recent developments, however, have amply demonstrated that this process can be, and is being, operated under conditions of strict control to give a uniform sound product of high economic value. It is clear that basic Bessemer steel has a field of wide general application, and while not providing steels suitable for certain specific purposes, for which other processes of manufacture are more suited, it will in its way provide certain types of steel which are better made by this process than by any other.

Monsieur **H. Malcor** (Institut de Recherches de la Sidérurgie, France) wrote : Of the steel produced in France, 70% is made in the basic Bessemer converter ; this paper by Dr. Dickie will, therefore, be read with much profit in my country.

It is very interesting to know that the phosphorus content can be pushed as far as 0·02% without over-oxidizing the metal, or at least while retaining it in such a condition that its state of oxidation can be brought back to that required to produce a steel giving a normal effervescence. Aluminium additions in the ladle are reputed to produce inclusions. It would be interesting to know the alumina contents of the steel ingots, or, better still, either the Fox count, or else some other index of the cleanness of the steel, and to know whether, in certain deep drawings these inclusions do not cause difficulties.

By proving that the use of two slags was not necessary, Dr. Dickie has greatly simplified the operation of the basic Bessemer plant, and has certainly diminished the difficulties of control.

The addition of mill scale instead of scrap is particularly interesting, especially when it is noted that the chilling effect of scale is of the same order of magnitude as that of scrap. This latter fact seems difficult to explain ; the specific heat of Fe_3O_4, between 0° C. and 1400° C. is about 244 kg.cal. per ton per ° C., hence the heat required to heat up one ton of dry scale to 1400° C. is 340,000 kg.cal. To this must be added the heat of reduction of Fe_3O_4 at 1400°, namely, 1,050,000 kg.cal.,* but the heat content of the blast economized deducted, namely, 457,000 kg.cal./ton of Fe_3O_4. One arrives, therefore, at :

933·000 kg.cal for scale, as against
210·000 kg.cal for scrap,

which corresponds to a ratio of 4·5, whereas experience at Corby shows that the true ratio is 1·5. This difference between calculation and experience is hard to explain. If it is assumed that the scale contains 10% of moisture, and that this is driven off at 100° C., the theoretical ratio reduces to 4·3, which is still much too high. During the discussion, Dr. Dickie stated that the experimental ratio of 1·5 could be confirmed by calculation ; it would be interesting to know this calculation.

Dr. Dickie's assertion that the iron content of the slag of ' improved ' steel casts was a little lower can be explained if these casts were systematically a little cooler than the others. I should like to know whether the temperature determinations had been sufficiently numerous to settle this question one way or the other.

I do not think that, in practice, variations in the blast pressure play any noteworthy part in the absorption of nitrogen by the bath. Some fifteen years ago, I showed that the pressure of the blast as it issued from the bottom was only of the same order as the ferrostatic pressure. The variations in this pressure with the depth

* Reckoned by assuming that the heat of formation of Fe_3O_4 at 0° C. is 266 kg.cal per gramme-molecule, and that the average specific heat of iron between 0° C. and 1400° C. is about 150 kg.cal per ton of iron per ° C.

of the bath are about 0·1 kg./sq. cm. at most, that is, less than 10% of the absolute pressure; it is hardly probable that these variations have any effect.

In France, we have already made some tests on blowing with blast containing 40% of oxygen; the resulting nitrogen contents were of the order of 0·004%. This reduction of the nitrogen content cannot be explained on the basis of the partial pressure of nitrogen; the explanation is probably furnished by Naeser's theory that steel does not absorb nitrogen so long as the bubbles contain oxygen. From this viewpoint a similar effect would be obtained by enlarging the diameter of the bubbles.

AUTHOR'S REPLY

Dr. **Dickie** wrote in reply: The results in the paper, as regards open-hearth steel, refer only to the best quality open-hearth. When I said that the Bessemer steel was not quite so good as the best quality open-hearth, what I really meant was that it is very much better than the worst quality open-hearth!

Mr. Honeyman mentioned the solubility of nitrogen in steel at 0° C. and gave the figure of 0·0075%. Actually there is a difference of opinion amongst investigators on this figure; other figures given for the solubility in α-iron at atmospheric temperature are 0·03%,[23] 0·015%,[24] 0·01 to 0·025%,[25] 0·001% or less,[26, 27] or 0·0004%,[28] but I am sure that if a steel is cold worked, the amount of nitrogen in solution will become very much less than the figure he mentions, and that the ageing changes are due in whole, or in part, to the lowering of the solubility by the cold deformation, followed by the separation of some form of precipitate, giving increased hardness and embrittlement.

Replying to Mr. Honeyman's query about the analyses of the steels used in the corrosion tests, results of which were given in Tables IV and V, these tests were done on the steels whose analyses are given in the last group of Table III.

I do not accept the comment that the cupping test appears to be an acceptance test. The test, as is indicated in the paper, was used as an exploratory test in the early stages of this work, when it was not possible to take scores or hundreds of casts of steel, and machine impact test-pieces from them for elaborate tests. It was purely an exploratory test, but it could conceivably be reduced to an acceptance test in particular cases, if sufficiently controlled.

Referring to brittle failure of mild steel without previous hardening, the alterations in composition which improve the strain-ageing properties also improve the steel in the other case; for example, as shown by notched tensile tests at sub-normal temperatures. The various steels described in the paper fall in the same general order of merit in each case.

Mr. Honeyman also asked about the effects of recarburizing the steels to, say, 30 tons/sq. in., and Mr. Thring also suggested that a factor for the carbon might be added to the Stromeyer index. It is true that increasing the carbon content has an effect somewhat similar to raising the phosphorus or nitrogen, but I should think that the fundamental reasons for this might be different in the case of carbon, which is always present in these steels in an amount higher than the solubility limit of carbon in ferrite.

In the paper, I am considering only very low-carbon steels. Raising the carbon content of the modified Bessemer steel to about 0·10 to 0·15%, whilst still retaining it in the unkilled condition, depreciates the impact-test results at low temperatures, but the results are still superior to the lower-carbon (and softer) ordinary Bessemer steel. Unkilled modified Bessemer steel with

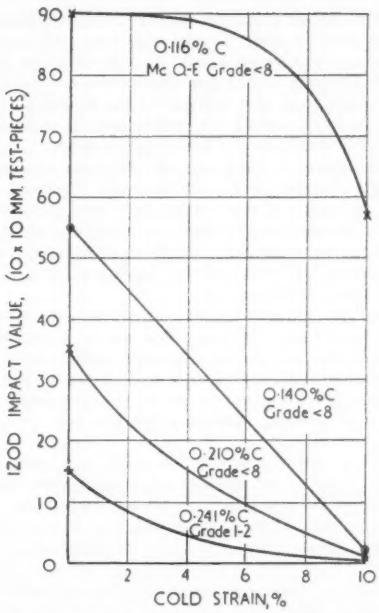


Fig. N—Impact value of killed Bessemer steels (German origin) at 20° C. (steels cold-strained, and aged at 200° C. for 2 hr.)

0·15% of carbon and 0·8% of manganese, for example, compares quite reasonably with open-hearth killed steels of similar composition.

Killed Bessemer steel with raised carbon content may show unusual effects. The accompanying Fig. N shows results obtained on killed Bessemer steels received from Germany. Impact tests were carried out after 10% straining and ageing. The steel represented by the top curve is quite good, and has a McQuaid-Ehn grain-size of less than 8. It also has, by analysis, sufficient aluminium in the steel to fix the nitrogen content, which is 0·010% in this case. The steel represented by the second curve from the top has a McQuaid-Ehn figure of less than 8, and a nitrogen content of 0·005%, but it contains insufficient aluminium to fix even that amount of nitrogen. In addition, it has a higher carbon content than normal Bessemer steel. This indicates that the presence of carbon is disadvantageous and is producing more embrittlement than would occur in a lower-carbon material. The other results in Fig. N, obtained on steels of still higher carbon content, confirm this conclusion.

The remarks of Mr. Honeyman, and others, on the effect of carbon content, draw attention to the much greater ease with which steels of very low carbon content can be made in the Bessemer process.

Dr. Macnair asked a specific question about the plugged steel. Our own experience is that we can make a blow of steel which is designed to go into plugged moulds for the controlled rimming steel, and that the same metal will rim with a generally level top in open moulds. In fact, our practice was really worked out with that in view, because it is sometimes not possible to be certain of having the necessary number of plugged moulds in every cast, and it is a great convenience to be able to make a good rimming ingot from the same ladle metal that is used for the plugged steel. Dr. Macnair mentioned that at Ebbw Vale they use a scrap process to control the temperature. They have, I think, an immersion thermocouple which they use on every blow. Our own experience is that temperature is not of great

importance: it is a factor, but it is quite possible to make the steel in such a way that one finishes with a good casting temperature and a low nitrogen content. Theoretically, one could get circumstances which were so favourable that the nitrogen was being evolved or expelled from the steel, in which case it would be an advantage to work at the high temperature.

Remarks made in reply to Dr. Macnair also apply to Dr. Leckie's query as to whether the effect of the ore or scale additions is due entirely to temperature. The scale is, of course, used in part to control the temperature in the same way as scrap, but its essential function is to provide direct oxidation of the bath, and to protect the air-blast from deoxidation.

Dr. Leckie asked whether we would use oxygen in addition to the ore or scale, and we shall certainly try that out in the near future. It may well happen that in due course we shall arrive at a still better process, using the oxygen partly as a control of the temperature conditions, and possibly allowing for the use of scrap, in addition to scale.

We do not dry the ore or the scale, but we take steps to see that they do not get into a bad condition. Generally speaking we do not like fine ore or scale; we prefer a certain amount of lumpy material, and it does not seem to matter whether there is some moisture in it, but it is better not left out in the open. No trouble has been experienced at Corby which could be attributed to the effects of hydrogen.

Reference has been made by Dr. Geller to skull in the ladle. The remarks that I have made on temperature cover that. There is no need to have unnecessary skulls in the ladle when using this new process. Naturally they do occur at times, but experience at Corby is that skulling is, on the average, only very slightly heavier than in ordinary practice.

We have bottom-poured the new steel on a number of occasions, in various sizes of ingots, and we have not found any difficulty due to low temperatures.

The number of tuyeres used is 270, evenly distributed across the bottom, and of 13 mm. dia.

Dr. Geller's notes on the effect of the age of the bottom in German practice are of interest. The experience at Corby is different from some of the experiences that he he quotes, and it is of a very definite character. In fact, repeated observations that a lower nitrogen was obtained in ordinary blowing when the lining was worn and the bottom new, probably gave the first clue to the ultimate procedure developed. Further statistical work at Corby, since the paper was written, indicates that it is more difficult to obtain a very low nitrogen content in the final steel when the nitrogen of the iron is high. The lower nitrogen contents of about 0·005% or 0·006% are generally associated with low nitrogen contents in the iron. Full account has been taken at Corby of the presence of special elements in the iron which fix the nitrogen in combination, and suitable methods of analysis have been developed to obtain correct estimations, and to distinguish between fixed and free nitrogen.

I have, in a general way, no reason to disagree with Dr. Geerlings' remarks. He showed that the new quality Bessemer steel, when processed to have a normal grain-size, is approaching the standard of the open-hearth steel which, I think he will agree, in his particular experiment, was supplied by us and was an aluminium-killed open-hearth steel. It may be that if we had supplied him with a silicon-killed steel his curve might have come below the Bessemer.

We have carried out a good many series of low-temperature impact tests at Corby, and we find that when the temperature control is sufficiently exact, little or no scatter is obtained.

The effect of grain-size, mentioned by Dr. Geerlings, has already been noted in the paper. The smaller grain-size naturally produced in the aluminium-killed open-hearth steels in equivalent treatments is, of course, one reason why it maintains a good impact value to lower temperatures than other steels. If the grain-sizes of aluminium-killed open-hearth steel and LNP Bessemer steel are reversed by appropriate hot- or cold-rolling and heat-treatment, so that the LNP has a fine grain-size, and the aluminium-killed open-hearth has a coarse grain-size, the impact curves at low temperatures are reversed; the killed steel then falls more rapidly in impact value than the LNP. This is illustrated in Fig. O.

The analysis of the example mentioned by Dr. Geerlings, which showed brittleness in spite of the nitrogen being low, may not be complete in this case. The high arsenic content of about 0·3%, in certain German steels, may have affected the results in the case mentioned, but other explanations are possible. I still maintain, with complete assurance in view of many confirmatory experiments, that when the nitrogen and phosphorus contents of the Bessemer steel are reduced to the same values as open-hearth steel, the differences between the steels, as regards susceptibility to embrittlement, disappear. Naturally the steels must be of the same type and have received the same treatment.

Mr. Robinson gave an interesting review of the position as a result of his enquiries into the subject of Bessemer steel. The effect of the shallow bath alone, with ordinary blowing and without scale, is to lower the nitrogen content from an average of 0·017% in the old form of bath, down to 0·014%, or 0·013%, in the shallow bath, with the usual 25 tons charge. It may well be that in converters of different shape the relationship would be different.

Mr. Glen gave the point of view of the practical steelmaker, and thanks are due to him for extending the information made available on the process, by his useful remarks on various practical aspects. He contributed considerably to this work himself, partly by being responsible for the development of the analytical methods when most of this work was done, and later by being in control of the steelworks at a vital stage.

He asked me two questions; one was with regard to the frequency curves in Fig. 19. I agree with him that these curves are representative of earlier results. They illustrate the position in 1943, and earlier; since that date there has been a progressive improvement in the frequency curves for what we call the ordinary Bessemer material, which is considerably better today than the former standard.

He also asked about the effect of notches on embrittlement. As mentioned in the paper, it is not necessary to have a notch for a piece of steel to split. We have applied X-ray stress determinations to the cupping

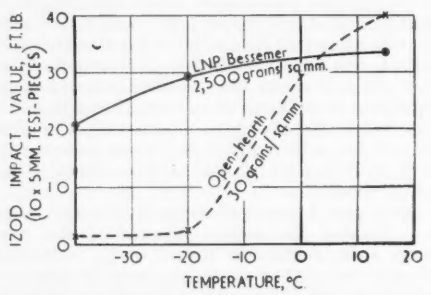


Fig. O—Izod impact values at low temperatures (steels in low-temperature annealed condition)

test-pieces mentioned in the paper and find that the internal stresses produced in the ordinary Bessemer steel are about 2 or 3 times higher than the stresses produced in the low-phosphorus, low-nitrogen steel. The existence of these internal stresses, besides the hardening of the steel, is evidently important in determining how the cupping test behaves.

I have already replied to Mr. Thring's remarks on the effects of carbon as far as we know them. He mentions the effect of surface area on the pick-up of nitrogen. We have done some work on laboratory models, and we have found that if the tuyeres have a funnel-shaped orifice, the air bubbles break up to a much greater extent than when the tuyeres are cylindrical, and give an increase in the nitrogen absorption. This is similar to the wear of the tuyeres in practice in certain cases, and illustrates the effect of surface area of contact between gas and metal. In side-blowing practice there is a smaller area of contact, and the metal surface becomes oxidized, thus enabling decarburization to proceed, and at the same time preventing, or minimizing, nitrogen absorption.

Replying to the contributors in the written discussion, I agree with Monsieur Dardel's remarks, with one or two reservations, but they do not alter the validity of the expression given by me for nitrogen absorption. I did not, of course, state that the solubility is proportional to T^5. The expression given in the paper is, as stated, for absorption in limited time periods, insufficiently long to reach saturation in blown metal, and typical of the time periods encountered in converter practice. The results of Chipman and Murphy, and of Kootz, quoted by Monsieur Dardel, and those of Eklund, already quoted by me, all agree that the variation in solubility with temperature is small. The largest variation, found by Kootz, is about 6% per 100° C. Variation in temperature has, however, a distinctly larger effect on the absorption in a limited time, as shown in the paper.

Any effect on the rate of absorption connected with change in composition or with the solubility at equilibrium, *i.e.*, with the 'pressure head,' $[N]_m - [N]$, will modify the diffusion coefficient in the expression given in the paper, in addition to the effects of surface films, as mentioned in the paper, and as found by Naeser.

I agree that the carbide or phosphide in pig-iron decreases the solubility of nitrogen, as shown by the investigators he mentions, and also by Eklund. They used, however, laboratory alloys, whereas in practice, at least at Corby, the solubility of nitrogen in pig-iron is much less than 0·009%. Most of the nitrogen is in combination as stable nitrides, which can be swept out by contact with soda slags.

The 'paradoxical' effect of the oxide additions mentioned in the paper, and referred to by Monsieur Perrin, is, I think, connected with the speedier completion of the reactions, the blowing time being shorter when using oxide additions. Accurate study of the slag conditions, throughout blowing, is not easy, owing to the non-uniform condition of the slag, except in the last stages. We make no alteration in the amount of lime used in the new process compared with the ordinary process.

Dr. Colclough's testimony, given in his remarks, is welcome at Corby, and we have to thank him for his views, which are based on extensive enquiries into Continental practice.

Replying to the point raised both by Dr. Geller and Monsieur Malcor about the relative cooling effects of scale and scrap, there is a relative loss of heat, as indicated in the paper, due to the absence of the heat of formation of oxide and to more heat being absorbed by the scale. Neither of these factors produces its full theoretical effect.

We have measured the amount of blast used in both scale blowing and ordinary blowing and find that in scale blowing the excess oxygen (blast plus scale), over that required for the bath reactions, is greater than the excess in ordinary blowing. There is, therefore, a greater loss of oxygen in scale blowing. This may be attributed partly to loss of scale blown out, and partly to the escape of unused oxygen.

Also, spectrographic examination of the converter flame indicates a shift in intensity towards the violet end of the spectrum, in scale (or ore) blows, compared with ordinary blows. This is in line with Dr. Geller's suggestion that more CO_2 is formed, with consequent higher flame temperature, which will tend to balance the greater theoretical cooling effect of the scale. These factors result in less absorption of heat by the scale and less net saving of heat of formation than would be supposed from the theoretical calculation given by Monsieur Malcor and mentioned by Dr. Geller. Whilst the broad reasons for the apparent discrepancies are thus explained, it is probably not possible to make a precise calculation, at least at this stage, owing to insufficient data on the various factors involved.

I do not think that the effect on the iron content of the slag, mentioned by Monsieur Malcor, is explained by temperature differences, although, clearly, variation in temperature will affect the amount of iron in the slag, in particular cases. The effect is still evident in blows which are of similar temperature, as judged by temperature determinations and by the presence or absence of ladle skulls. Probably much more has still to be learned about the factors involved in this aspect.

Replying further to Monsieur Malcor, the alumina contents of the ingots produced are 0·006% to 0·011%, by analysis of slab samples. We have not carried out Fox inclusion counts, which are probably not really applicable to this type of steel, but the steel is approximately of the same cleanness as ordinary Bessemer steel. The inclusions in Bessemer steel, whilst relatively large in number, are of small size and evenly distributed in the structure; they do not become rolled out in an objectionable way. Experience in making stampings, and the like, from cold-rolled strip has not revealed any injurious effects arising from such inclusions as are present. The new steel has been successfully used, in bulk quantities, for several applications which previously required the best quality deep-drawing open-hearth steel.

If we accept Monsieur Malcor's finding that the blast pressure when leaving the tuyeres is just above the ferrostatic pressure, the pressure will still be proportional to the bath depth, as well as being affected by other factors which decrease the partial pressure of nitrogen, as explained in the paper. The importance of pressure in practice can be directly appreciated from the increase in nitrogen absorption which occurs when the nose of the converter becomes unduly constricted, causing back-pressure.

The point about large bubbles, mentioned by Monsieur Malcor, is of interest, and I agree with his conclusion. We find at Corby that the use of tuyeres of larger diameter than usual (while maintaining the same total area) produces a slight lowering of the nitrogen content.

Finally, I have to thank all those who have taken part in the verbal and written discussions for the valuable additions they have made to our knowledge and appreciation of this subject. I also desire to testify again to the excellent assistance and co-operation received from all concerned at Corby during this development.

References

1. E. W. Ehn: *Metal Progress*, 1931, vol. 19, May, p. 89.
2. W. Böddeker: *Mitteilungen aus den Forschungsanstalten von G.H.H. (Gutehoffnungshütte) Konzern*, 1939, vol. 7, pp. 145–146.

3. W. GELLER and A. WILMS : *Stahl und Eisen*, 1941, vol. 61, Pt. I, pp. 337–349.
4. T. KOOTZ : *Verein deutscher Eisenhüttenleute*, 1944, *Vertraulicher Bericht, No.* 72.
5. K. DAEVES, R. FRERICH, and W. HOLTMANN : *Ibid.*, p. 5 (Discussion)
6. D. R. Patentanmeldung N. 46. 381 VIa/18b and N. 46, 961 VIc/18b (inventor : J. Klärding). NPN-Steel of the Neunkircher Eisenwerke A.G. with Low Phosphorus and Nitrogen Content.
7. G. NAESER and W. BADING : *Verein deutscher Eisenhüttenleute*, 1944, *Vertraulicher Bericht No.* 67.
8. T. KOOTZ and G. GILLE : *Stahl und Eisen*, 1948, vol. 68, pp. 287–294.
9. H. WENTRUP and O. REIF : *Ibid.*, 1944, vol. 64, pp. 349–358.
10. K. DAEVES and W. HOLTMANN : *Verein deutscher Eisenhüttenleute*, 1944, *Vertraulicher Bericht No.* 62.
11. C. J. SMITHELLS : " Gas and Metals." London, 1937, Chapman and Hall.
12. R. A. FOWLER and C. J. SMITHELLS : *Proceedings of the Royal Society*, 1937, vol. 160, Series A, pp. 37–47.
13. J. CHIPMAN and D. W. MURPHY : *Transactions of the American Institute of Mining and Metallurgical Engineers*, 1935, vol. 116, pp. 179–196.
14. T. KOOTZ : *Archiv für das Eisenhüttenwesen*, 1941, vol. 15, No. 42, pp. 77–86.
15. Y. DARDEL : *Metal Progress*, 1947, vol. 52, No. 2, Aug., pp. 252–256.
16. Y. Dardel : *Circulaire d'Informations Techniques du Centre de Documentation Sidérurgique*, 1948, vol. 5, pp. 33–42.
17. J. C. VAUGHAN and J. CHIPMAN : *Transactions of the American Institute of Mining and Metallurgical Engineers*, 1940, vol. 140, pp. 224–232.
18. M. M. KARNAOUKHOV and A. N. MOROZOV : *Comptes Rendus de l'Academie des Sciences, U.R.S.S.* (*Doklady*), 1947, June, pp. 735–747.
19. G. NAESER : mentioned in a paper by W. OELSEN : *Stahl und Eisen*, 1948, vol. 68, p. 186.
20. R. VON SETH : *Jernkontorets Annaler*, 1924, vol. 79, pp. 1–108.
21. Deutsch Brassert Gesellschaft : French Patent No. 897,505.
22. Y. DARDEL : French Patent No. 943,054.
23. C. B. SAWYER : *Transactions of the American Institute of Mining and Metallurgical Engineers*, 1923, vol. 69, p. 798.
24. A. FRY : *Stahl und Eisen*, 1923, vol. 43, 4th Oct., p. 1271.
25. O. EISENHUT and E. KAUPP : *Zeitschrift für Electrochemie*, 1930, vol. 36, p. 392–404.
26. W. KÖSTER : *Archiv für das Eisenhüttenwesen*, 1930, vol. 3, pp. 637–655.
27. L. J. DIJKASTRA : *Phillips Research Reports*, 1947, vol. 2, p. 357.
28. J. D. FAST : *Phillips Technical Review*, 1948, vol. 10, No. 1, p. 26.

Discussion on the Paper—
INGOT STRUCTURE IN A SERIES FROM RIMMED TO KILLED STEELS MADE FROM THE SAME CAST* *by P. M. Macnair*

Dr. **P. M. Macnair** introduced his paper.

Dr. **C. H. Desch,** F.R.S. (Past President, London) : This paper represents the culmination of a good deal of work done in the Ingots Committee. After much discussion in the earlier meetings on the theory of the solidification of ingots we have come back to the principles enunciated some time ago, which are now found to be sufficient. If we take into consideration the differential freezing produced by the solidification of the upper part of the ingots before the reactions in the interior are completed, I think that the peculiarities of these structures are fully accounted for. That seems obvious now, but in the earlier days of the Ingots Committee there was a good deal of controversy on these points, and certain members of the Committee would not accept the principle of differential freezing as sufficient.

I find myself in agreement with what the author has said, so that I have no detailed comments to make, and, with the photographs which, of course, were very fully discussed by the members of the Ingots Committee, I think that we now have quite a logical account of the reactions which take place and of the structural changes which occur in the solidification of ingots of this type of steel.

Dr. **D. Binnie** (Lancashire Steel Corporation) : I notice, on passing from ingots 1 to 5, the direction of increasing aluminium additions to the ingot mould on teeming, that the depth of the rim remains substantially the same, and also, as the author has emphasized, the purity of the rim remains unchanged. From an examination of Tables II to V and especially for sulphur and manganese contents, I would go farther and say that the rim of the ingots with higher aluminium addition is rather more pure as the aluminium addition is increased. There must not be a great deal of free oxygen in these aluminium-treated rimming and rising steels, and yet there is the full rimming action giving a rim even more pure than if no aluminium addition had been made at all. The view has been expressed many times that the amount of free oxygen necessary for the rimming action need only be small. The amount of gas given off seems to be excess gas without any major effect on the rimming action. Whiteley has dealt with this point.†

I refer now to the ingot shown in Fig. 24, which was teemed when the mould was at a tilt of 20°. There is no modification whatever in the depth of the rim, and, apart from the top level of the freezing steel, there is no indication that the ingot had been tilted. An air gap, probably of double the usual dimensions, would only be on the upper vertical wall, and yet the rim depth is precisely the same along the wall with an air gap and the wall without an air gap. This proves that apart from early capping of the steel, or closing the top of the mould by an addition of, say, aluminium to the top of the ingot, the rimming action will proceed normally, and some drastic alteration in practice is necessary to alter its depth.

In Figs. 4 to 7 the author has taken his average analysis for the rim and core by adding them together and dividing by 2. It can be seen from measurement that the volume of the rim is about twice the volume of the core, so that the average is more correctly taken by twice the rim, plus the core, divided by three. If that is done then the curve for average carbon in Fig. 4 does not rise steadily from ingot 1 to ingot 4, but follows more closely the contour

* *Journal of The Iron and Steel Institute*, 1948, vol. 160, Oct., pp. 151–163.

† *Journal of The Iron and Steel Institute*, 1942, vol. 146, No. II, p. 371P.

of the curve shown for the rim, which shows that the change from rimming to semi-skilled and killed steel is much more sharp than Fig. 4 would suggest.

The author would have liked an ingot intermediate to those shown in Figs. 5 and 6. Theoretically, of course, this intermediate ingot could have been obtained, but I think the conditions would be rather critical. These balanced steels are to be associated more with a steel higher in carbon and in manganese than the author has used, when the freezing of the steel becomes more under control.

I am in general agreement with the theories expressed in the paper, but, of course, the value of the author's work will be augmented when the examination on selected sections is complete. I would like very much to see a detailed analysis across the rim of ingots 1 and 4 to find what exactly is the effect of the aluminium additions on the purity of the rim.

Dr. **H. A. Dickie** (Stewarts and Lloyds, Ltd., Corby): The author's results are complementary to work published a few years ago by Mr. Mitchell, in which steel with a constant oxygen content, and also a constant carbon content, was cast in moulds of different sizes. Mr. Mitchell showed that the smallest ingot had the largest ratio of rim to core, and that this ratio decreased as the ingot size was increased. It is also known that increasing the carbon content, which also decreases the oxygen content, acts in a similar way to the deoxidation, without raising the carbon, as carried out by Dr. Macnair. The rim becomes more shallow and the core wider. We now know, thanks to the work done by the author and some of the previous work that I have mentioned, the effect of a fair range of variables in altering the structure of rimming and semi-rimming steel.

There are no general objections that I would take to the author's theory of solidification, but I am a little surprised that in discussions which have taken place on this subject in Committee, attention has not apparently been drawn to the difference between making additions to the mould and making additions to the ladle. It is clear from the author's results that he has produced a killed steel in the mould with considerably smaller additions of aluminium than would be necessary to kill the steel in the ladle and to produce the same, or a very similar, structure in the cast ingot.

I have had some experience of killing steel in the mould. It goes back a good many years, and it happened at Corby, but there is no one now at Corby, except myself, who had anything to do with it in the original stages, and I do not accept any responsibility for it. The practice there was to hold a shovelful of aluminium pellets on the edge of the mould and shake them into the mould while the steel was being poured; killed steel was produced with approximately $\frac{1}{2}$ lb. of aluminium per ton of steel, whereas the same steel killed in the ladle would require 4 lb. of aluminium per ton of steel. Complaints came in from users, and we altered the practice, and supplied rimming steel with slightly raised carbon. The steel had to be 30–32 tons tensile. We supplied it on trial to the people who had complained, and in every case they were perfectly satisfied that the trouble had been rectified. The rimming steel was much superior in a variety of uses to the steel which had been killed in the mould.

We then wrote to various other people who were getting this steel, pointing out that we were changing the quality to a much better quality, and we hoped that this would be all right. One of the largest users, however, wrote back that we must on no account change the quality, because the previous quality had been satisfactory in every respect. It just shows what may happen in different usages of steel. I think it should be borne in mind, in any further examination of the subject by the Committee, that the oxygen relationship in these ingots cannot be the same as in steel that is killed in the ladle.

Dr. **L. Reeve** (Appleby-Frodingham Steel Co.): There are two points which I wish to raise. The first is that the change-over from ingot No. 5 to ingot No. 6, from the "rising" to what Dr. Macnair calls the "semi-killed," is so sudden. By a very slight change in the deoxidizer, he has produced what is a balanced ingot, or almost so. The terms "semi-killed," "slightly worked," "balanced" and so on, are sometimes used as synonyms, but at what stage would Dr. Macnair himself use the word "balanced" to describe the ingot concerned? The point has some importance in view of the very large tonnage of balanced steel which is made in this country, and as, in the summary to the paper, the author says that no true balanced ingot was obtained in this series, I should like to hear his comments.

The second point is in connection with the oxygen determinations which are to be made; were oxygen samples taken in the bath just before tapping? If not, it is a great pity, because in some work I am doing it appears that the oxygen content of the steel at tapping, despite subsequent deoxidation, has a significant influence on the thickness of the skin in the type of balanced ingots which I shall discuss in a paper which I am presenting at this meeting.*

Sir **Charles Goodeve**, F.R.S. (Director, British Iron and Steel Research Association): While I agree generally with the theoretical considerations in this interesting paper, I think our general physical/chemical knowledge of reactions in liquids allows us to go further in visualizing the mechanism of the carbon/oxygen reaction. From our knowledge of the thermal motion of the atoms or molecules, the frequency with which carbon and oxygen atoms collide can be calculated, and for a concentration of both of 0·05% there are about 10^{30} of such collisions per cubic centimetre per second. Every collision does not necessarily lead to a reaction, but only those with which is associated sufficient 'activation' energy, E_A, to bring the atoms over the energy barrier separating the old from the new configuration. This fact leads to the well-known $e^{-E_A/RT}$ term in reaction kinetics. For reactions involving atoms the values of E_A are found to be small and, even if we allow for some substantial bond formation between the iron atoms and the carbon or oxygen, it is difficult to imagine a value of E_A for the carbon/oxygen reaction exceeding, say, 25 kg.cal. Indeed, even if we put in a value of twice this, we only bring the 10^{30} down to 10^{24} (equal to two gram-molecules) as the number of successful collisions per cubic centimetre per second. This means that the rate of the reaction would be two gram-molecules/c.c./sec., which, to all intents and purposes, is instantaneous. At all times, therefore, there must be an equilibrium between carbon, oxygen, and dissolved carbon monoxide.

Why, then, does not the carbon monoxide escape very rapidly? It does, of course, at a surface, but in the case considered here the formation of bubbles is the principal mechanism. The first condition for the formation of a bubble is that the concentration of carbon monoxide must exceed that given by equilibrium with a gas phase at the prevailing pressure. The formation of a nucleus of a bubble in a homogeneous liquid is a notoriously difficult process; it is difficult to conceive in theory, and in fact it is doubtful if it has ever been observed. The energy necessary to make a small hole against the

* *Journal of The Iron and Steel Institute*, 1948, vol. 160, Oct., pp. 169–176.

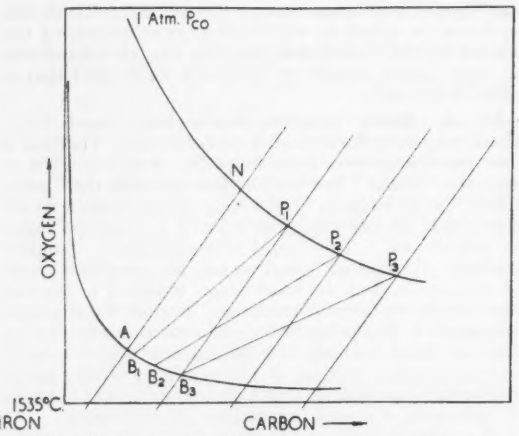


Fig. *A*—Iron/carbon/oxygen ternary diagram

surface tension of liquid metal is very large. Most gas evolution occurs at the surface of particles of foreign matter where a break or hole has already been made in the liquid, or at the surface of the container. This general experience undoubtedly applies to metals, although in the case under consideration the interface between the liquid and solid phases may also play a part in providing a mechanism for bubble formation.

One of the main characteristics of gas-evolution processes is that extensive supersaturation frequently occurs, especially in systems very free from foreign matter. We already know that supersaturation up to a factor of two commonly occurs in the open-hearth furnace.* It seems likely therefore that it will play an important part in the processes considered in this paper.

Mr. **H. Brooke Freeman** (John Summers and Sons, Ltd.): I am particularly interested in the mechanism of rimming and the formation of the rim zone. Reference is frequently made to " skin thickness " but there appears to be some confusion between skin thickness and rim-zone thickness. As steelmakers we are interested in the thickness of the zone, but we are more interested in the thickness of the solid skin which forms in the early stages of freezing and which protects the rim holes from exposure in the soaking pits.

I have long viewed the rim zone, at about mid-height of the ingot, as consisting of five different layers. The first layer is formed of small pellets of steel sprayed from the surface of the metal rising in the mould ; on this sprayed layer a " chilled layer " forms, and this in turn is followed by the formation of the third layer which may be considered as the " solid skin." According to Hultgren and Phragmén this solid skin is formed under such rapid conditions of CO formation that the gas bubbles formed on the freezing face enlarge so quickly that they break away from the face and avoid entrapment. As evidence of their temporary existence they leave behind " rim channels " in the solid skin. As freezing proceeds the solid skin of the third layer merges into a fourth layer, which contains numerous rim holes, and this in turn is followed by the fifth layer, in which rim holes are no longer formed but are substituted by rim channels.

According to Hultgren and Phragmén the rim holes form when the CO production is only sufficient to produce bubbles which grow at the same rate as the freezing face ; they are thus entrapped. Therefore the third, fourth, and fifth layers freeze in a sequence of conditions implying firstly, rapid CO production, and secondly, a slower CO production, followed by a rapid CO production, all relative to a unit weight of the steel frozen.

The production of CO relative to a unit weight of metal freezing may now be considered from the point of view of the iron/carbon/oxygen ternary diagram (Fig. *A*). It is accepted that as freezing proceeds the composition of the molten liquor tends to move along NP_3, and that of the frozen metal along AB_3. Maximum CO production occurs per unit weight of steel frozen from a steel at composition N. Steels to the right of N, of compositions lying on NP_3, produce less CO per unit weight of steel frozen as the steel composition moves towards P_3. Thus a steel of composition, say, P_1, which produces enough CO in its early stages of freezing to give a solid skin, may, as the composition of the molten liquor passes along to, say, P_2, only produce sufficient CO to form rim holes. If then, the molten-liquor composition moves to P_3, with a further decrease in CO relative to unit weight frozen, how can the cessation of rim-hole formation, and its replacement by a solid zone containing rim channels, be explained ?

Dr. **P. T. Carter** (Royal Technical College, Glasgow) wrote : I should like to ask Dr. Macnair how far he has considered the interpretation of his results in the light of the well-known theory advanced by Hultgren and Phragmén* in 1938. Their interpretation of the effect of pressure by combining CO + CO_2 equilibrium isobars with the liquidus of the Fe/C/O system is more fundamental, and would obviate the introduction of incorrect phraseology (*e.g.*, last sentence, p. 160). It seems better to visualize gas evolution as being opposed by the combined atmospheric and ferrostatic pressure, especially as in some cases it is completely suppressed, when the " back reaction " rate will be zero. The true back reaction rate is determined by the concentration of carbon monoxide dissolved as such in liquid steel. A carbon-monoxide solubility has already been postulated by Chipman and his co-workers to explain the variation in the [C] [O] product at constant temperature and pressure. The saturation concentration of carbon monoxide for a given temperature should be proportional to the applied pressure, and must be attained before gas evolution can theoretically commence.

Hultgren and Phragmén also introduce the concept of the balanced composition. Can Dr. Macnair explain the variations in carbon and oxygen found by Swinden,† without postulating the existence of such a composition ?

From the amount of aluminium required to produce the killed steel (equivalent to 0·040% of oxygen) it is estimated that ingot No. 2 (no addition) contained about 0·035% of oxygen before teeming. This figure may be subject to correction after oxygen analyses have been completed, but it cannot be far out. On this basis, points representing ingots 0–9 are plotted on the Fe/C/O diagram in Fig. *B*, together with isobars for one and two atmospheres. (The pressure in the ingots during free gas evolution varies from one atmosphere at the top, to two atmospheres at the bottom.) It can be seen that ingots Nos. 0, 1, and 2 should all show gas evolution before freezing, (*i.e.*, during teeming) which would result in a lower ingot weight, as observed. Ingots 3–7 should show

* S. Fornander, " The Behaviour of Oxygen in Liquid Steel during the Refining Period in the Basic Open-Hearth Furnace " and subsequent discussion by F. D Richardson : Faraday Society Discussion, No. 4 on " The Physical Chemistry of Process Metallurgy," pp. 296 and 335 (1948).

* A. Hultgren and G. Phragmén, *Jernkontorets Annaler*, 1938, vol. 122, No. 8, pp. 377–465.

† T. Swinden, *Journal of The Iron and Steel Institute*, 1942, No. II, pp. 345P–356P.

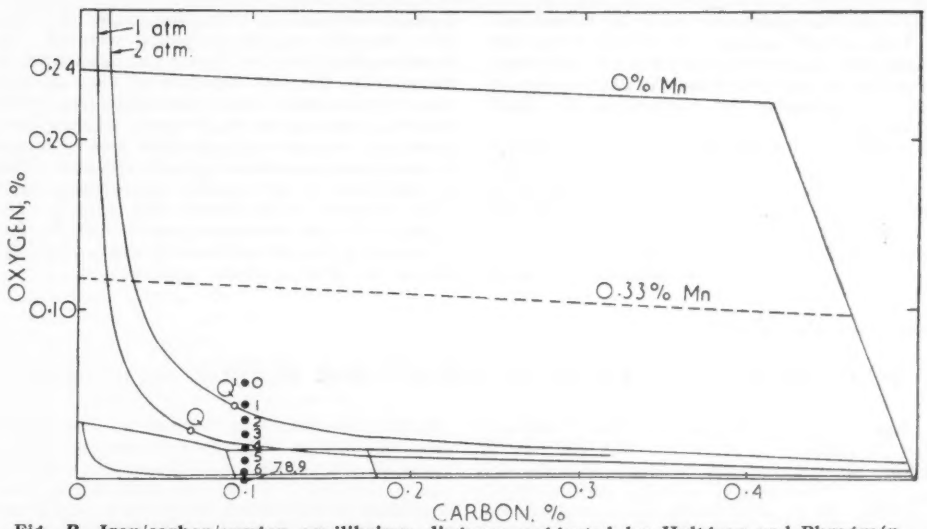


Fig. *B*—Iron/carbon/oxygen equilibrium diagram suggested by Hultgren and Phragmén. Isobars drawn using data of Chipman and co-workers. Q and Q′ are the balanced compositions for 1 and 2 atmospheres and 0·33% Mn calculated by Binnie

progressively decreasing rates of gas evolution once freezing begins, because of: (*a*) The smaller amount of gas evolved—the ratio of gas evolved to solid formed can be obtained by applying the lever principle to Hultgren and Phragmén's diagram; (*b*) the increasing range of temperature, corresponding to freezing, over which gas evolution takes place. For both reasons the balanced composition should show the most rapid rate of gas evolution.

Finally, it is noted that with regard to primary blowhole formation in the lower part of the ingot, No. 1 is out of line, showing more primary blowholes than No. 0 and No. 2. Can Dr. Macnair give any reason for this?

AUTHOR'S WRITTEN REPLY

Dr. **P. M. Macnair** wrote: The contributions of the various speakers have been most encouraging, and I would like to repeat Dr. Desch's statement that the paper may be considered to represent the culmination of a good deal of work done in the Ingots Committee, to which the industry was so much indebted for previous papers. Dr. Binnie has made several interesting suggestions for further work, which, I feel, are desirable. The ingots have been retained and are available to anyone who desires portions for particular examination.

Dr. Dickie has pointed out the probable difference between the effect of a ladle addition and a mould addition. It was felt that the difficulties arising from an experiment controlling ladle additions far outweighed any possible advantages that might accrue from slavishly following steelworks practice, and the idea was rejected for the present series. If it was considered desirable to obtain the difference in effect (which would probably be most marked on non-metallic inclusions) on any type of ingot which had been made from mould-treated metal, as opposed to ladle-treated metal, or even furnace-treated metal, with no ladle or mould addition, then comparison with the present series might be helpful in elucidating this variable. At least one plant in America is now making rimming steel in the ladle, and converting it to killed, and so-called stabilized, steel in the mould by an aluminium addition of 4 lb./ton.

Dr. Dickie further suggests that the oxygen relationship in these ingots cannot be the same as in steel killed in the ladle. This is probably true so far as total oxygen is concerned, but not necessarily so in the case of free oxygen, and it is free oxygen which controls the structure of the resultant ingot. It is not a simple matter to determine free oxygen, but at least a first approximation can be obtained by splitting total oxygen into FeO, MnO, SiO_2, Al_2O_3, etc.

The sudden change between ingots Nos. 5 and 6, mentioned by Dr. Reeve, was unexpected, and, to that extent, remarkable. Dr. Reeve rightly mentions that it adds further evidence to the extremely critical nature of the oxygen content required to make balanced steels. I feel that the internal structure of a balanced steel is not clearly established. Regarding his own opinion of the structure, it is felt that it would certainly contain no pipe or shrinkage cavity, but instead, top central blowholes such as those shown in ingot No. 6. It would have no other central or secondary blowholes. There is no satisfactory evidence to show the extent of primary blowholes. There may be a few, but only a few, and these all at the top end of the ingot. It may well be that the terms "semi-killed" and "balanced" are really synonymous, in which case there would be no primary blowholes.

It is easy to agree with Dr. Reeve that it it is a pity that no oxygen samples were taken in the bath, or for that matter in the ladle or during teeming. A great many variables have still to be separated and investigated before a complete picture is available.

The contribution from Sir Charles Goodeve was very interesting in helping to show fundamental ideas of the mechanism of the carbon/oxygen reaction. I was particularly pleased to have support in the statement: "At all times . . . there must be an equilibrium between carbon, oxygen, and dissolved carbon monoxide" (in the liquid metal phase). It was my belief in this statement that caused me to reject the idea that the CO evolved during the setting of a rimming ingot arose from change in equilibrium conditions caused by falling temperature in the still-liquid steel. But the CO is evolved, and on p. 158 it is suggested that this is perhaps best accounted for by the upsetting of the "instantaneous" equilibrium of the metal phase, when a new solid phase appears on solidification of the ingot in the mould.

I am in general agreement with Mr. Freeman's remarks. It is difficult to accept the idea that the fifth zone is produced by a fast evolution of CO, and apparently Mr. Freeman has the same impression. Perhaps he has overlooked the suggestion put forward in the fourth paragraph on p. 161.

Several extremely interesting, even if provocative, ideas are put forward by Dr. Carter. Unfortunately space is not available to go into them with the detail that they deserve. I have looked again at the last sentence on p. 160, since it was suggested that I was using incorrect phraseology. The sentence still seems to convey the meaning that was intended, although I admit that the exact meaning might have been more fully brought out by additional wording. It does, however, seem that Dr. Carter has fallen into the same difficulty in the next sentence of his communication, when he states that " the back reaction rate will be zero." Reaction rate can never be zero, whether forward or backward, except at the absolute zero of temperature. It always has an actual value, however small. When there is equilibrium, it is a kinetic, not a static, equilibrium. With reference to Dr. Carter's last point I am not very happy about the structure shown in ingot No. 2. It may be caused by the difficulties of burring over which arise during the planing of the section.

Discussion on the Paper —

INGOT SURFACE DEFECTS IN STRUCTURAL STEELS* *by L. Reeve*

Dr. **L. Reeve** (Appleby-Frodingham Steel Company) introduced the paper.

Dr. **P. M. Macnair** (Messrs. Richard Thomas and Baldwins, Ltd.) : The problem of ingot surface defects is a major one. In its worst phase it means discard of material which has been melted and rolled. In an intermediate phase it may mean only degrading the finished product. At its best it means some recovery, but only after expense has been incurred in scarfing or chipping.

Balanced steels, which the author discusses, are formed at a stage where the setting ingot is very sensitive to small changes in free-oxygen content. With too little oxygen the semi-killed type of ingot is produced, with a slight excess a rising type is made. It is probable that a balanced ingot contains a few primary blow-holes at the top end, and these extend further down as free oxide increases. The author thought that a primary cause of surface defects was the extent to which primary blow-holes were formed, and more particularly the actual skin depth. It may well be that the various factors mentioned by the author—mould design, etc.—influenced the formation of primaries, and this would then be the main reason for potential surface defects.

The author added 4 oz. per ton of aluminium at the top of the ingot during teeming. Both the quantity and the time of the addition are important. Such a large addition would certainly kill the steel at the top end and thus produce no primaries, but it would have little effect lower down. That is where the time factor is so important, as may be illustrated from the equation mentioned by the author, $d = KT^{\frac{1}{2}}$.

The author states that the first 1 in. of skin is produced in 1 min., and that in 4 min. the skin is 2 in. deep. But if the equation is used in a decreasing time factor it means a $\frac{1}{2}$-in. skin in $\frac{1}{4}$ min., $\frac{1}{4}$-in. in $\frac{1}{16}$ min., and $\frac{1}{8}$-in. in $\frac{1}{64}$ min., or roughly 1 sec. That is the important point ; 1 sec. after the teeming has left any level, the thickness of the solid skin is $\frac{1}{8}$ in., and the blow-hole formation has started, as this is a typical skin thickness in balanced and rising steels. There may be some doubt about the actual value of K immediately freezing sets in, but there seems to be considerable evidence from this equation that primary blow-hole formation does start lower down the ingot before the late aluminium addition has been put into the mould. Has the author had any experience with his corner sampling technique in finding primaries in the bottom half or just above it, and none at the top ?

In discussing ingot surface defects, the author avoids the effect of relative size of section by multiplying by 2 when the section is only half the size, and so on proportionately. I agree with that. It has been said that defects can be removed by extra rolling ; that may be so, but it is perhaps not the whole story. If a piece is rolled to a standard length and a standard thickness there may be only one flaw present ; if now it is rolled to twice the standard length and half the standard thickness and is then cut to two standard lengths, that flaw will appear in only one piece. In order therefore to make a comparison it is necessary to allow for the relative thickness considered. I would like the author to comment on that, as I have found similar circumstances in dealing with sheets and tinplate, where defects are proportionately fewer as the section is reduced.

Mr. **N. H. Bacon** (Messrs. Steel, Peech, and Tozer) : This question of balanced steel is often a source of misunderstanding, and if I may say a word or two it may help to clear it up. Balanced steel can be made in two ways : (1) By reducing the degree of oxidation of the steel by ladle additions, usually 0·05-0·15% of silicon, according to the particular steelmaking method employed, but insufficient to kill the steel fully, so that a late generation of gas will occur during the process of solidification to form enough blow-holes to offset largely the pipe cavity ; (2) by teeming normal effervescing steel into the moulds and making an aluminium addition of 4–6 oz. per ton to the steel when the mould is almost full. This deoxidizes the top part of the metal so that effervescence ceases immediately and the top of the ingot freezes over. The normal primary blow-holes which have already begun to form are " nipped in the bud." The latter is the type of balanced steel dealt with by Dr. Reeve, and the method used is, I think, the better one for many purposes. It is a method often used at Templeborough, with the additional advantage of bottom pouring. Bottom pouring ensures slower teeming, which results in a thicker skin of solid steel.

Dr. Reeve's top, bottom, and middle corner samples show the characteristic distribution of blow-holes associated with effervescing steel, but they are only partially developed, owing to the artificial suppression of the effervescence by the late aluminium addition.

On the subject of primary blow-hole formation, Dr. Macnair, in his paper† on ingot structures in a series of steels, put forward the hypothesis that the solid steel often to be found in rimming-steel ingots between the primary and secondary blow-holes, is due to the temporary cessation of the carbon–FeO reaction caused by back pressure from the freezing over of the top surface of the

* *Journal of The Iron and Steel Institute*, 1948, vol. 160, Oct., pp. 169–176.

†*Journal of The Iron and Steel Institute*, 1948, vol. 160, Oct., p. 161.

ingot. I think a simpler explanation is that the primary blow-holes continue to grow in conjunction with the advancing wall of freezing steel, the rate of advance slows down, and the still-liquid metal loses its heat at a much slower rate ; thus conditions arise which permit the gas still being formed to escape freely.

This total escape of the gas continues until the top becomes frozen over, when the gas still being generated is trapped *in situ*, so forming the characteristic system of secondary blow-holes which always occurs at the junction of the rim and core zones.

Dr. Reeve raises a very important question in referring to the effect of sulphur on cracking. Sulphur is becoming a bigger problem than ever, because our raw materials are getting worse. Lime, which contained 0·10% of sulphur before the war, now contains 0·25%, and fluorspar, instead of containing 0·2%, now often carries 0·7%. Iron ores, too, are increasing in sulphur, and there is a general increase throughout in the sulphur content of the steel. Since this affects the cracking tendencies of steel, it is a matter of great importance.

There has been some controversy on this question, whether it is the sulphur itself which makes the steel more prone to crack, or whether the conditions in the furnace resulting in that high sulphur are the primary cause of the cracking. The author implies (though he does not definitely say so) that in his opinion it is the sulphur itself which is the cause of the cracking, and in that I agree with him, because a good deal of evidence is coming to the Ingot Surface Defects Sub-Committee in support of that view. One argument in favour of it is that the sulphur in the final steel is really no true reflection of the working of the charge. One cast may finish with 0·05% of sulphur and another with 0·03%, with exactly similar conditions of working and slag basicity ; the only reason for the difference is that the sulphur charged into the furnace in one case was only about half the amount charged into the furnace in the other case.

Some melting shops in this country are fortunate in that their sulphur burden is about half that of others. The result is that they can work with a much smaller slag volume and still finish with a lower sulphur. This brings up the important point that every effort should be made to reduce the sulphur burden to the steelmaker. It could be done if sufficient attention was given to it, *e.g.*, by using better quality fuel in calcining the limestone, instead of using outcrop coal, as is often done, with 2–2½% of sulphur. Fluorspar could be treated to remove the sulphur ; it might cost twice as much per ton but would be well worth it. Again, there are well-proved methods of reducing the sulphur content of pig iron. These are not practised by some ironmakers because of the very real difficulties in the way, but every effort will no doubt be made to overcome such difficulties.

The author suggests that cool steel gives a thicker solid skin than hot steel, but I believe that this apparent result is due only to the fact that cool steel cuts out the ladle nozzle to a less extent, resulting in a slower teeming speed, and that it is the slower teeming speed that is the cause of the thicker skin. Or, put in another way, if two casts of effervescing steel were teemed at similar speeds, the only difference being the temperature, then the hotter cast would give the thicker skin, owing to the easier passage and escape of the gas through the more fluid steel.

AUTHOR'S REPLY

Dr. **Reeve** replied : Dr. Macnair points out that a true " balanced " ingot contains a few primary blow-holes in the top end only. Most so-called " balanced " ingots are usually slightly more oxidized, and will have an ingot skin thickness of ⅛ in. or more, with primary blow-hole formation present at both the bottom and the middle, which may be suppressed by late aluminium additions in the upper part of the ingot. Dr. Macnair's inquiry as to whether I have observed this state of affairs is covered by Fig. 13 of my paper, where it will be noted that the top section is almost, though not entirely, free from such blow-holes. This ingot had an aluminium addition to the top of the mould.

The extension of the equation $d = KT^{\frac{1}{2}}$ to cover times as short as 1 sec. is not justified by the experimental technique used in the determination of the rate of solidification of ingots. As most of you are no doubt aware, this technique involves emptying the contents of the ingot mould at short time intervals after completion of teeming, and it is obvious that with such a technique it would be impossible to fill and empty an ingot mould in times as short as those discussed by Dr. Macnair. However, I am willing to grant him the probability that at the end of a few seconds the thickness of the solidified skin may be between ⅛ and ¼ in., and that late aluminium additions can influence blow-hole formation in the upper levels only of the ingot. In this connection, I should like to emphasize the point—which is already mentioned in my paper—that in discussing the rate of growth of ingot wall thickness (in the first part of the paper) I have not considered the possible influence of the primary blow-holes. As pointed out on page 171, the influence of such blow-holes on the rate of solidification of the ingot has not been investigated, but possibly some delaying action will occur. The discussion of ingot skin thickness and its influence on sub-surface defects, which appears in the second part of the paper, is quite independent of the discussion on the rate of growth of the ingot wall in the first part.

Dr. Macnair's confirmation of my method of correcting for the influence of section weight on defectives is welcomed. I think there are actually two causes for the direct relationship between ingot defects and section weight ; the first is the one which Dr. Macnair discussed, *viz.*, the fact that, when working to a standard length, the loss of one of these lengths, owing to a defect, must mean a smaller weight lost when the section thickness is small. The other reason is that if semi-finished material is reheated before being rolled into finished material, a certain amount of ' washing ' of the surface occurs, which is more likely to remove sub-surface defects the lighter the section.

Mr. Bacon's remarks about the importance of sulphur will be supported by all who have practical experience of the influence of this element on ingot cracking. His point about the relationship between cooler steel and increased teeming rate is correct. I doubt, however, whether this is the sole reason for the apparent relationship between ingot skin thickness and decreasing temperature ; I think there is a specific temperature effect. The matter may be worthy of further investigation.

Mr. Bacon points out quite correctly that there are two ways of making balanced steels, and he states that in my paper I was dealing with the type of balanced steel produced by the addition to an ' effervescing ' steel of 4 oz. of aluminium per ton when the mould is almost full. It is true that I have referred to this method in the earlier part of the paper, but, for the sake of accuracy, it should be pointed out that actually both methods are used at our works. The method described is applied generally to the smaller ingots used for rolled sections. In the case of plates, depending upon their thickness, a combination of the two methods mentioned by Mr. Bacon is normally employed, *i.e.*, an addition of some silico-manganese is made to the ladle, and a small aluminium or alsimin addition to the mould top.

Correspondence on the Papers—

THE FORMATION OF SIGMA PHASE IN DUPLEX CHROMIUM-NICKEL-MOLYBDENUM CORROSION-RESISTING STEELS*

by H. W. Kirkby and J. I. Morley

and—THE OCCURRENCE AND SOME EFFECTS OF SIGMA PHASE IN A MOLYBDENUM-BEARING CHROMIUM-NICKEL AUSTENITIC STEEL*

by L. Smith and K. W. J. Bowen

Mr. **H. J. Goldschmidt** (The B.S.A. Group Research Centre, Sheffield) wrote: I should like to congratulate Kirkby and Morley on their excellent paper, which, together with that of Smith and Bowen, throws much valuable light on sigma-phase formation. It might perhaps have been more desirable if not one, but several steels of varying composition had been investigated together, to ascertain how far the observed trends are generally significant; and a reference to phase-diagram relations might have assisted in explaining some of the observations.

I was very interested in the authors' X-ray examinations, both by diffraction and by micro-radiography, and would like to ask whether a fundamental difficulty has not been found in detecting and quantitatively estimating the ratio sigma/ferrite/austenite, because the result depends so much on the method of surface preparation. This is demonstrated conclusively by the two useful etchants given, one dissolving sigma, the other the matrix; but in our experience any such preferential attack gives a misleading answer to grazing-incidence photographs, as reflections from the recessed phase are screened by that in relief. Would the authors consider that a particular mixture of both reagents offers a uniform removal of material? Some etch is necessary because a straight cold-worked (unetched) surface is liable to give erroneous results, as it is known, for instance, from Cook and Jones' work, that mechanical work has a profound effect on sigma formation. This aspect is not mentioned in the paper, and I wonder whether during specimen preparation (sectioning, polishing) special precautions were thought necessary to eliminate the effect.

The most important conclusion, in my view, is that sigma may be formed even for a chromium content of the steel well below that expected, because sigma precipitates out of a parent ferrite which is richer in chromium than the austenite. During investigations on sigma-phase formation in heat-resisting steels we encountered a manner of producing this phase which I believe has not been pointed out before. This is by decarburizing a carbide. It has, in common with Kirkby and Morley's mode of sigma formation from ferrite, the fact that in each case the parent phase is one abnormally enriched in chromium, while the nominal chromium content of the steel would not have suggested its formation. The effect is illustrated by the following:

A carbide extract was prepared from a heat-resisting steel of 13/13 Cr–Ni base, containing 2·4% of tungsten and 0·44% of carbon. X-ray tests showed this to consist of the carbide $Cr_{23}C_6$ with Fe and W in solution; the cube edge (a_0 = 10·623 kX.) and chemical analysis indicated a Cr/Fe ratio within the carbide $(Cr, Fe, W)_{23}C_6$ of approximately 3/2. This carbide was then heated *in vacuo*, with, however, a residual air pressure of 10^{-4} mm. Hg. On X-raying again, the product revealed itself to be mainly sigma. Some chromium-rich ferrite, as well as some initial carbide was also present. Decarburization had evidently taken place as follows:

$$(Cr, Fe, W)_{23}C_6 + nO_2 \longrightarrow FeCr\ (\sigma) + Cr, Fe, W\ (\text{ferrite}) + (CO_2 + CO)$$

Emphasis is laid not so much on these particular experimental conditions as on the principle underlying the reaction, *i.e.*, that sigma phase can be produced from carbide, which in practice may occur in a variety of ways. Thus, even for a steel relatively low in chromium, which would appear to lie well below the danger region, sigma formation can, in fact, take place if a secondary phase (in this case the carbide) assumes a Cr/Fe atom ratio favourable to sigma (order of 1/1, but with considerable latitude), and if, for any reason, carbon is extracted from this carbide. Several such reasons may be foreseen in practice, of which the following are instances:

(1) On combustion-engine parts running, for example, at 700–850° C. and exposed to gaseous attack. (2) By acid attack or through other corrosive liquors. (3) By the presence of a stronger carbide-former such as Ti or Mo, rendering the Cr carbide unstable and freeing the Cr for sigma formation. (The latter might be termed "internal" decarburization, as opposed to "external" decarburization in the case of (1) and (2).)

Thus the advantages of adding stronger carbide-formers in certain conditions may be offset by a liability to sigma formation, unless special precautions are taken in deciding the alloy composition in the light of service conditions. A remedy would be (*a*) to introduce, by added elements, a primary carbide more stable than $Cr_{23}C_6$, but such that Cr is also soluble in it and not free to form sigma; (*b*) to encourage the formation of Cr_7C_3 rather than $Cr_{23}C_6$; or (*c*) if $Cr_{23}C_6$ is formed, to adjust the ratio Cr/(Fe + Co + Ni) so as not to be near unity.

Smith and Bowen, in their paper, mention that for corrosion resistance both carbide and sigma formation should be considered simultaneously; they do not, however, pursue the subject. The carbon content of the steel used by them, as well as that used by Kirkby and Morley, was, however, so low that these considerations will be less important, but, in principle, I think they should not be neglected.

With regard to the paper by Smith and Bowen, I have read this with great interest. Some of my notes on Kirkby and Morley's preceding paper on sigma phase would equally apply to theirs, so perhaps I may be allowed to refer to these for any comments the authors may care to make.

* *Journal of The Iron and Steel Institute*, 1948, vol. 158, Mar., pp. 289–305.

Table A

INTERPLANAR SPACINGS OF Σ-PHASE

Anderson and Jette (converted) (CrK_α)	Smith and Bowen	Goldschmidt (CoK_α)
3·24 m	...	...
3·10 w	...	...
2·97 w	...	...
2·505 w	...	...
2·37 vw	...	...
2·29 s	...	...
2·26 w	2·279 vvw	2·26 vw
2·12 vs	2·124 vw	2·14 s
2·06 s	...	2·075 m
2·03 s	...	2·01 m
1·97 vs	1·973 mvw	1·97 ms
1·93 vs	1·928 mw	1·925 s
1·88 vs	1·884 w	1·88 m
1·84 m	1·839 vw	1·84 w
1·762 w	...	1·76 vw
1·637 vw	...	...
1·389 vw	1·392 vw	...
1·372 vw	...	...
1·349 vw	...	...
1·324 w	...	1·33 w
1·254 vs	1·259 Trace	1·259 w
...	1·241 m	1·241 m
1·229 m	1·230 vw	1·228 w
1·226 m	...	...
1·218 sm	...	...
1·202 sm	1·208 mw	1·208 mw
...	1·200 vvw	...
1·183 m	1·189 vvw	1·188 vw
1·173 s	...	1·172 m
1·162 vs	...	...
....	...	1·138 m
...	...	1·071 m
...	...	1·039 m
	...	1·005 ms
...	...	0·999 wm
...	...	0·991 wm
...	...	0·985 ms
...	...	0·962 wm
...	...	0·945 m

vs = very strong. s = strong. ms = medium/strong. m = medium. wm = weak/medium. w = weak. vw = very weak.

Table B

INTERPLANAR SPACINGS OF Σ-PHASE

Smith and Bowen	Goldschmidt (CoK_α)	Kirkby and Morley (CoK_α)
...	...	2·38 vw
...	...	2·31 wm
2·279 vvw	2·26 vw	...
2·124 vw	2·14 s	2·14 s
...	2·075 m	2·07 ms
...	2·01 m	2·03 ms
1·973 mvw	1·97 ms	1·98 s
1·928 mw	1·925 s	1·94 s
1·884 w	1·88 m	1·89 ms
1·839 vw	1·84 w	1·86 vw
...	1·76 vw	1·78 vw
1·392 vw	...	1·375 vw
...	1·33 w	1·335 vw
1·259 Trace	1·259 w	1·260 m
...	...	1·252 m
1·241 m	1·241 m	1·241 m
1·230 vw	1·228 w	...
1·208 mw	1·208 mw	1·205 w
1·200 vvw	...	...
1·189 vvw	1·188 vw	...
...	1·172 m	1·171 vw
...	1·138 m	1·146 w
...	1·071 m	1·065 m
...	1·039 m	1·040 m
...	1·005 ms	1·011 m
...	0·999 wm	...
...	0·991 wm	0·989 w
...	0·985 ms	...
...	0·962 wm	0·966 vw
...	0·945 m	0·950 w
...	...	0·936 vw

vs = very strong. s = strong. ms = medium/strong. m = medium. wm = weak/medium. w = weak. vw = very weak.

In the section on X-ray examination, the authors give the interplanar spacings of sigma, saying that no previous table of spacings could be found in the literature. It may interest them that a table has, in fact, been published by Anderson and Jette.*

In Table A, the American authors' results (converted to interplanar spacings, kX. units, from the $\sin^2\theta$ values quoted) are compared with Smith and Bowen's, and also with some of my own. Disparities in the results need clarification.

AUTHORS' REPLIES

Mr. **H. W. Kirkby** and Mr. **J. I. Morley** (Brown-Firth Research Laboratories, Sheffield) wrote: We thank Mr. Goldschmidt for his kind remarks and for his interesting contribution to the discussion of our paper. We agree that the examination of a range of steels would have been more desirable and the paper improved by reference to phase diagrams, had these been available. However, we look forward to making further contributions on this subject in relation to other steels.

With regard to the difficulty of making a precise estimate by X-ray methods of the ratio sigma/ferrite/austenite in a steel specimen, Mr. Goldschmidt's experience has evidently been similar to our own. The mode of etching before X-ray examination certainly can have a great effect on the results, depending on the resistance to chemical attack of the various phases present. In the paper, as will have been observed, we did not attempt to make any quantitative estimation of the proportion of sigma phase by X-ray methods, but relied on the general picture given by magnetic tests, mechanical properties, and micro-examination. The etching reagents referred to in the paper were chosen for their selective attack on the sigma phase or on the matrix, and were used only in the micro-examination. For the X-ray diffraction studies, where the detection of the sigma phase was the main object, a different reagent was used, as it was necessary to avoid the selective pitting of constituents. The reagent used for this purpose was 10% hydrochloric acid, which gives a gentler and more uniform attack. Other reagents of similar action were also available, but some degree of relief is probably inevitable. As regards the mixing of the two reagents, suggested as a means of obtaining uniform attack, whilst we have not attempted this, we doubt whether such an attack could be obtained. Perhaps the best solution, in some cases, to the problem of removing cold-work effects from the surface without unduly emphasizing relief effects, is a good electrolytic polishing technique. Occasionally, however, constituent relief may be helpful, as when relatively small quantities of a phase are to be identified by the grazing-incidence method. In such cases the screening of the matrix by the unknown phase in relief serves to increase the relative strength of its reflections and so assists identification.

The reference to cold-worked surfaces is a little puzzling, as in our experience the only important effect of cold work on sigma formation is revealed by subsequent heat-treatment, and this we believe was the effect referred to in Cook and Jones' work. Hence in this case the estimation of the ratio sigma/ferrite/austenite by X-ray methods on a cold-worked untreated surface would be influenced only indirectly as the result of ferrite

* *Transactions of the American Society for Metals*, 1936, vol. 24, p. 375.

formation from austenite during the cold working, and then only in cases where the austenite is unstable.

The formation of sigma phase from complex carbides under circumstances such as external decarburization is not only interesting, but quite novel. It is a phenomenon which in certain cases might need careful attention, but, as Mr. Goldschmidt suggests, in the low-carbon steels with which this paper was concerned, the extent of any such effect would not be of great importance.

We should like to ask Mr. Goldschmidt (*a*) whether he can be certain that his carbide extraction is complete, *i.e.*, with no residual steel, since we feel that this would affect the results, and (*b*) whether he has succeeded in obtaining sigma phase from carbides in a solid steel specimen, where the relative rates of diffusion of carbon and chromium would assume great importance. We have, however, no actual experience of sigma formation from carbides.

With regard to Mr. Goldschmidt's comments on Smith and Bowen's paper, we are able to record the interplanar spacings of the sigma phase extracted as a residue from our molybdenum-bearing 18/8 steel (*see* Table B). These are only approximate, to within $\theta = \frac{1}{2}°$, as the method used was the rapid one described by D. W. Davison.* However, the agreement is sufficiently good to leave no doubt about the identification of the sigma phase in relation to other published work.

Mr. **L. Smith** and Mr. **K. W. J. Bowen** (Imperial Chemical Industries, Ltd.) wrote in reply : The carbon content of the steel used in our experiments was typical of commercially supplied steels of this type. However, we did not make any reference to the different rôles of carbide and sigma on the corrosion resistance of the austenitic steels, except by implication. We think that these rôles are quite distinct, and have read reports in which there has been obvious confusion between the effects of sigma formation and carbide precipitation.

We are grateful to Mr. Goldschmidt for drawing our attention to the paper by Anderson and Jette, and for his own list of interplanar spacings. We cannot, however, agree that there are any significant or inexplicable discrepancies. Generally speaking, the lines we obtained are weaker than those quoted by Anderson and Jette and by Goldschmidt, and we report fewer of them. It must be borne in mind that we were working with more complex alloys, and with a structure relatively weak in sigma. In addition, at least one of the lines (2·01) quoted by him, and not by us, suggests that it is an alpha line and not a sigma line. The fact that we seem to cover only the middle of the field, while Anderson and Jette give higher, and Goldschmidt lower, figures than ours, is probably due to the known difficulty of detecting lines at the edges of the field.

With regard to Mr. Goldschmidt's first remarks in his contribution to the paper by Kirby and Morley, we only say that both of his points are of great significance, but the work he suggests would take years to carry out. Actually an attempt is already being made by the Alloy Steels Research Committee, at the National Physical Laboratory, to develop phase-diagram relations for these types of steels.

* *Journal of Scientific Instruments*, 1948, vol. 25, No. 1, Jan., pp. 7–10.

Correspondence on the Paper—

THE FLUIDITY OF STEEL*

by R. Jackson, D. Knowles, T. H. Middleham, and R. J. Sarjant

Dr. **W. Ruff** (Novara, Italy) wrote : The exact and very valuable investigations of the above authors have brought the study of the fluidity of steel to a stage where it seems almost as if a fairly long pause would not be out of place. There are, however, still many problems to be solved, which, to some extent, take us back to 1936, when a fluidity test was designed using a very narrow channel, which, however, was straight, as distinct from the usual spiral form. It was pointed out at the time, that the purpose of the tests was not the invention of a fluidity test but was the establishment of the general laws which determine the fluidity of steel and of malleable cast iron ; another type of test was shown to be much more useful from a scientific point of view, an outflow test, in which the rate of flow of the steel leaving the mould was measured. The fact that this second type of test was ignored for years does not affect the value of tests in which the metal solidifies after running for a certain time in the mould ; nevertheless, the exclusive employment of solidification tests leads to a very one-sided assessment of the fluidity, which can largely be avoided by using the outflow test of the second type, in the investigation of the constant conditions of flow.

The writer has, himself, neglected this outflow test in the intervening period, primarily because he concentrated on two questions which constantly recur in steelworks practice : How can steel be tested at the furnace, immediately before tapping, and on a works scale, with maximum reliability, and how can such results be used ? The authors, in their paper in 1947, contradict what R. J. Sarjant and T. H. Middleham wrote in 1938, and give the impression that they now advise against using the " straight channel test " with a channel of small diameter, and that they prefer the old " spiral test," although it is admitted that the spiral test suffers from numerous defects which can sometimes not be corrected in time, on isolated results. Objective evaluation of a larger number of results, on the other hand, did not succeed in proving that the straight-channel test is no longer of practical value. For example : According to the authors, a spiral test, 29 in. long, requires at least 57 c.c. of steel, excluding the " knots " on the spiral, and a larger gate, against 7·7 c.c. of steel for a corresponding straight-channel test, 17·0 in. long, *i.e.*, at least 7·4 times as much. For this reason, the spiral test can hardly be carried out in routine steelworks production, immediately before casting from large furnaces ; the sampling spoon is sufficient to take the small volume required for the straight-channel test out of the furnace, and in the evaluation of a large number of data a simple

**Journal of the Iron and Steel Institute*, 1947, vol. 157, pp. 1–21.

procedure using a sampling spoon has for many years proved itself completely practicable. It is obvious that a test which depends on the amount of heat introduced—that is, on the weight or the volume of the substance examined—will produce smaller errors with larger than with smaller volumes, when the differences in the amount of heat lost in unit time exceed the fluidity effects which can still be observed. The straight channel test has a considerable advantage over the spiral test owing to the reduced heat losses. The amount of heat lost by a unit volume depends on the surface, and in this respect the surface of the spiral test is related to that of the straight-channel test in the ratio 3·8 to 1, in a test giving a run of 29 and 17 in., respectively, in the two forms of mould; this implies that although the amount of heat introduced in the spiral test is 7·4 times as great as that in the straight-channel test, the heat losses are also increased, so that a factor of only 4 is left for the probability of error in the straight test. The above authors have observed this probability of error factor of 4 in a comparison of the straight-channel and spiral tests, but it does not follow that when such errors have been eliminated by extrapolation that the new values for the straight-channel tests thus obtained are less significant than spiral-test results. The present writer, in fact, assumes the exact opposite, namely, that the spiral test obscures the true fluidity conditions.

If, bearing this in mind, we now return to the position reached in 1936, we shall probably be forced to admit that the most suitable method of eliminating the cooling effect of the surface as a factor influencing fluidity is still the outflow test described at that time, for with the aid of this test it was at least possible to reduce the question of fluidity to the problem of flow at constant temperature, the laws of which are incorporated in the $R\lambda$ diagram. The relation of the viscosity of the steel to the temperature was assumed to be known, and the temperature factor was thus eliminated. But what does this mean to the steelworks operator, for whom it remains a fact that acid steel is more fluid than basic steel? It corresponds to a steeper "flow of steel" curve for acid, than for basic, steel on the $R\lambda$ diagram; in other words acid steels, owing to their lower viscosity ν, have a greater tendency to change from laminar flow to turbulent flow, and as a result of this turbulence, acid steels may, under certain circumstances, solidify relatively earlier if a large cooling surface is available. If, however, the cooling surface during the test is not so great, the lower viscosity of the acid steels becomes apparent, and the straight-channel tests are relatively longer than those of basic steels. In spite of the lower viscosity of acid steel it is quite possible for it to give shorter final lengths than the less fluid basic steel, if the influence of the heat losses through the surface is very significant, as in the spiral test. The exact conditions are difficult to predict accurately, but straight-channel tests have never yet indicated such a complete reversal.

Increased resistance to flow causes a change from laminar flow to turbulent flow, as can be observed on the $R\lambda$ diagram, corresponding to the rate of flow v; this also explains why a steel begins to solidify at points where its turbulence produces the maximum cooling rate, because such a point in the channel of flow represents a disturbance in the equilibrium of heat input and heat loss, and the increased heat loss accelerates the advance of solidifying crystals from the wall of the channel to the centre, so that the flow of the steel may even be interrupted, while the liquid steel, beyond the point of disturbance, runs on. In such cases in spiral tests, very fluid metals may form tongues, up to 5 cm. long, without the steel completely filling the cross-section; in straight-channel tests such points of disturbance are more difficult to find, because, when the sample is broken up, they appear only in the form of small holes in the centre of the section, which are sometimes confused with blow-holes. The authors have apparently also observed such slightly puzzling tongues in spiral tests on their particularly fluid manganese steel No. 4.

To facilitate discussion of the authors' valuable experimental results, some of their results are collected in the appended Table A, from which conclusions other than those reached by the authors may well be drawn.

The doubtful character of the advantages of the spiral test over the straight-channel test, as has been pointed out with special reference to differences in the surface areas and the elimination of scatter by extrapolation, is probably due to the number of straight-channel tests being insufficient for a fair comparison. In the case of melt 2, where the mould was too short owing to the great fluidity of the steel, the straight channel might well have been lengthened, or, alternatively, the diameter have been reduced. The figures in Table A should show whether the fluidity is indicated more correctly in the straight-channel or in the spiral test.

According to Table A the acid steel F432, is more fluid than the basic steel F295, in the straight-channel test, but less fluid in the spiral test.

Furthermore, according to Table A, the straight-channel test indicates that steel 3 is more fluid than the basic steel F295, whereas this result is again reversed in the spiral test. This is particularly noticeable in the case of the 15-in. straight-channel test result at immersion temperatures of 1650° and 1633° C., and the corresponding optical temperatures.

Table A

COMPARISON OF THE BEHAVIOUR OF DIFFERENT LIQUID STEELS

Steel No.	Alloying Constituent	Heat No.	Furnace Lining	Immersion Temperature, °C	Optical Temperature, °C	Spiral Test, Length, in.	Straight-Channel Test Length, in.
1	CuSi	F295	basic	1600	1505	24·0	12·5
,,	,,	,,	,,	1650	1545	29·0	15·9
1	CuSi	F432	acid	1600	1490	22·0	13·3
,,	,,	,,	,,	1650	1540	29·0	17·0
3	Low C Low Si	F319	basic	1600	1495	22·0	12·5
,,	,,	,,	,,	1650	1540	29·0	16·6
,,	,,	,,	,,	1633	1525	26·0	15·0
2	SiNi	F312/858	basic	1600	1495	29·5	>25·0
,,	,,	,,	,,	1550	1450	19·5	—
4	Mn, 13%	F324	basic	1600	1500	32·5	18·0
,,	,,	,,	,,	1550	1450	26·5	15·0

According to Table A, again, the silicon–nickel steel 2 is more fluid in the straight-channel test than is the manganese steel 4, and in this case, too, these relations are reversed in the spiral test. All these reversals in the spiral test are due entirely to more rapid cooling of the steel during the test, as a result of the difference in surface areas between the spiral and the straight-channel tests. This effect is particularly striking in the comparison of steels 2 and 4, since the fact that the samples of steel 2 always ran out of the straight-channel moulds shows that they could certainly have exceeded 25 in., so that we are entering a range of values where straight-channel and spiral tests gradually reach the same length ; this means that a straight channel with a diameter of $\frac{3}{16}$ in. gives the same length of run as a spiral channel with a cross-section of 77·4 sq. mm., corresponding approximately to a diameter of $\frac{3}{8}$ in. The writer envisages considerable future development of steels such as steel 2. It is therefore not the degree of superheating which causes the increased fluidity, but rather the different conditions at the wall of the channel of flow, whether these be due to a different type of turbulence on the part of the steel, to the prior solidification of crystals of a different nature, or only to variations in the relation between temperature and viscosity. In this connection, investigation of the influence of nickel on the fluidity of steel is particularly interesting ; unfortunately, up to the present time the means are lacking for the study of these questions with the aid of the outflow test described in 1936.

AUTHORS' REPLY

Dr. **Jackson,** Mr. **Knowles,** Mr. **Middleham,** and Professor **Sarjant,** wrote in reply :

Much of Dr. Ruff's contribution arises from a misconception of the significance of the term "fluidity." In our paper, this was used as a convenient term to indicate the capability of a steel to be cast, so that a more fluid steel will be one which can be cast in longer, thinner, or more complicated types of mould.

By reducing the question of fluidity to the problem of flow at constant temperature, Dr. Ruff is assuming that we are dealing with fluidity in its true physical sense as the reciprocal of viscosity, and not, as we actually used it, as another term for castability.

In his comments, Dr. Ruff calls attention to the apparently contradictory conclusions in our 1938 and 1947 reports. The objects of the two papers were dissimilar, the first being an investigation of the Ruff straight-channel type of test mould, and the second designed to explain apparent discrepancies found in comparing the temperature/fluidity relationships in our earlier work with those published by Taylor, Rominski, and Briggs. The so-called contradiction which he points out is actually strong evidence in favour of our opinion, since the later tests were designed to overcome some of the admitted shortcomings of the earlier tests.

The results of our more recent work show that the spiral test is subject to much less error than is the straight-channel test. We did not, on this account, dismiss the straight-channel test as of no practical value, but, on the evidence of the results of carefully controlled tests, we concluded that the spiral test was definitely superior.

The question of the convenience of the test for use under normal melting-shop conditions is one with which we were not primarily concerned, but no insuperable difficulty would be anticipated in using the spiral test under such circumstances. The volume of steel involved, though larger than is necessary for the Ruff test, would still be easily capable of being dealt with in a suitable sampling spoon.

In comparing the relative fluidities of acid and basic steels, the fallacy of the generally accepted opinion in foundry circles that acid steel is more fluid than basic steel was explained on the basis of the relative emissivities of the two types.

Dr. Ruff's selection of isolated points from amongst a mass of data, as in Table A, is fundamentally wrong. The aim in our experiments was to collect sufficient data for random errors to be eliminated. An argument based on selected observations, and ignoring all others, is untenable, and the conclusions Dr. Ruff has drawn regarding the relative fluidities of acid and basic steels are directly contrary to the indications of the series of tests as a whole.

The theoretical arguments by which Dr. Ruff compares the two types of test mould are interesting and ingenious, though some are not quite clear, for example, his consideration of heat content and heat loss in the moulds, which apparently ignores the fact that though the heat loss is proportional to the diameter the heat content varies according to the square of the diameter.

The suggestion, that the tongue on the tip of the test-piece is due to liquid in the centre being extruded through a small opening in an otherwise solid mass of metal, is not accepted. A more satisfactory explanation is based on the theory of surface tension, and an oxide film which has to be pushed along by the following metal.

In the test-mould the conditions of flow are laminar rather than turbulent, except at the point of entry into the channel, and here the conditions of turbulence will be more pronounced in the straight-channel type of mould. This argument also raises the further possibility that apparent castability may vary not only with the steel but also with the type of runner which is used for any particular casting. This opens up a fresh field entirely outside the scope of our paper, and one in which we should not be justified in entering in this reply.

It is surely preferable to judge the reliability of our work on the basis of the success, or otherwise, in actual casting, than to discuss it in the theoretical manner adopted by Dr. Ruff.

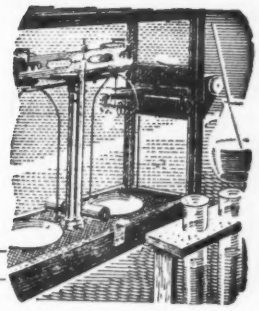
We maintain that we have demonstrated the value of the spiral test as a practical test to determine castability. The outflow method suggested by Dr. Ruff clearly bears no relation to actual casting practice, except in so far as the casting is run through by means of a head overflow.

METAL GRINDING IN FACTORIES: PROPOSED AMENDING REGULATIONS

The Minister of Labour and National Service has given notice that he proposes to make two codes of Special Regulations, amending, respectively, the grinding of metals (Miscellaneous Industries) Regulations, 1925, and the Grinding of Cutlery and Edge Tools Regulations, 1925.

The proposed Regulations are the outcome of the decision of the Court of Appeal in the case of Franklin *v.* Gramophone Co., Ltd. (1948 I.K.B. 542), in which certain difficulties arose as to the interpretation of the Grinding of Metals (Miscellaneous Industries) Regulations, 1925. In the light of these difficulties the Minister undertook to review the Regulations, and he has reviewed at the same time the Grinding of Cutlery and Edge Tools Regulations, 1925, regarding which similar difficulties are liable to arise.

B. I. S. R. A.



BRITISH IRON AND STEEL RESEARCH ASSOCIATION

Reviews of Certain Aspects of Metal Physics

FOREWORD

When the Metallurgy Division of B.I.S.R.A. was established in 1945, it was realized that no existing body under the aegis of the Association could appropriately include research on such subjects as gaseous and metallic diffusion, the fundamental strength of metals, crystal structures and their deformation in cold work, precipitation-hardening processes, and internal stresses. A Metal Physics Committee was accordingly appointed with the following terms of reference:

> "To encourage the study of the physics of metals such as theoretical strength, thermal and electrical conductivity, diffusion processes, magnetic properties, etc."

The Committee is concerned with the more fundamental work; practical applications are dealt with by other Committees of the Association. The members were selected from leading scientists, and arranged into Groups. These Groups, by informal discussion, assess the problems which require investigation and which are not already being studied. They include crystal structure, diffusion in alloys, phase precipitation and co-operative phenomena; mechanical properties; magnetic properties and analysis; electrical and thermal properties; gas–metal systems and surface phenomena. The constitution is as follows:

Mr. D. A. Oliver (*Chairman*)	B.S.A. Group Research Centre
Professor E. N. da C. Andrade, F.R.S.	University College, London
Mr. G. L. Bailey	British Non-Ferrous Metals Research Association
Professor Sir Lawrence Bragg, O.B.E., M.C., F.R.S.	Cavendish Laboratory, Cambridge
Dr. Bruce Chalmers*	Atomic Energy Research Establishment
Dr. Ezer Griffiths, F.R.S.	National Physical Laboratory
Mr. G. T. Harris	Messrs. William Jessop and Sons, Ltd.
Dr. W. Hume-Rothery, F.R.S.	Inorganic Chemistry Laboratory, Oxford University
Professor N. F. Mott, F.R.S.	H. H. Wills Physical Laboratory, Bristol University
Dr. E. Orowan, F.R.S.	Cavendish Laboratory, Cambridge
Dr. C. E. Ransley	British Aluminium Co., Ltd.
Dr. G. V. Raynor	Metallurgy Department, Birmingham University
Mr. T. F. Russell	English Steel Corporation, Ltd.
Mr. J. W. Ryde, F.R.S.	General Electric Co., Ltd.
Dr. C. J. Smithells, M.C.*	British Aluminium Co., Ltd.
Professor W. Sucksmith, F.R.S.	Physics Department, Sheffield University
Dr. H. Sutton	Ministry of Supply
Dr. C. Sykes, F.R.S.	Brown-Firth Research Laboratories
Professor A. M. Tyndall, F.R.S.*	H. H. Wills Physical Laboratory, Bristol University

(* Denotes original member, now resigned)

The function of the Committee and its Groups can be divided into four stages:

(1) To review current work in the respective fields.
(2) To encourage further study of metal physics, particularly in those sections where weakness is apparent, and to make appropriate recommendations for support.
(3) To co-ordinate research activity in the whole field.
(4) To organize private discussion or to initiate public discussions.

Much of the fundamental knowledge governing iron and steel can be obtained more easily by a study of some other metal or alloy system possessing fewer complicating factors. For this reason the Committee does not restrict itself merely to a study of iron and steel.

One of the first steps which were taken was the preparation of reviews of selected fields of research. These have surveyed the important work carried out during the war years, and while not claiming to be exhaustive nevertheless provide a reliable basis in planning new work. It is intended to publish in the *Journal* from time to time these valuable critical surveys, and to make reprints freely available to University and industrial research workers. Application should be made to B.I.S.R.A. following publication of a specific survey.

Part I—X-RAY WORK IN SOME FIELDS OF RESEARCH ON METALS AND ALLOYS, 1939 to 1946

By Audrey M. B. Douglas, M.A., Ph.D.

SYNOPSIS

This review covers the following subjects :

(*a*) Crystal structure determination of elements, alloys with hydrogen, carbon, and nitrogen, and intermetallic phases. (*b*) Lattice parameter determination. (*c*) Phase diagrams of binary, ternary, and quaternary systems. (*d*) Transformations in alloys, including age-hardening, dissociation in copper–iron–nickel alloys, phase-change mechanisms, and decomposition of copper–manganese alloys. (*e*) Growth, deformation, and recrystallization textures in metals. (*f*) New techniques and apparatus, including cameras, X-ray tubes, Geiger counters, monochromators, low-angle scattering measurements, study of diffuse reflections, micro-radiography, study of surface perfection, accuracy of lattice parameter determinations, and the " bubble " model for metals. Brief reference is also made to neutron diffraction.

CRYSTAL STRUCTURE DETERMINATION

Elements

THE structures of the metallic elements are, in general, simple and well known ; hence researches in this field have been few in number. In a re-investigation of thallium,[1] the structure was found to be body-centred cubic above 230° C. ; there are extra lines on the X-ray powder photograph which are not accounted for, and further investigation is required.

A preliminary study of the two allotropes of polonium[2] showed that the transition temperature was 705° C. An X-ray investigation showed that precipitated tellurium was not amorphous, as had been supposed, but that it had a hexagonal structure[3] ; high-temperature studies did not support the hypothesis that two allotropes exist. X-ray studies have been made of the amorphous, monoclinic, and hexagonal forms of selenium, and of the rate of transformation of the two former to the latter structure.[4] Other studies of amorphous selenium and also of explosive antimony showed that both have a liquid-type structure with co-ordination number and atomic distances very similar to those of the crystalline forms.[5]

There have also been some researches into the imperfections in the structure of the elements. Cobalt does not seem to obey the phase rule in that two phases, face-centred cubic and close-packed hexagonal, appear to exist in equilibrium at room temperature. No satisfactory general explanation of this has been advanced, though it is probably due to a small difference of free energy between the two phases ; this may also be the explanation for the imperfections present in the hexagonal phase which result in the broadening of some of the X-ray reflections. Both hexagonal and face-centred cubic structures can be regarded as built up of layers of close-packed spheres, though in different sequences ; in cobalt the sequence is usually of the hexagonal type, but occasionally it changes to cubic.[6, 7] The degree of imperfection varies from specimen to specimen, and no general rules for its occurrence and magnitude have yet been formulated.

Tin has been investigated by Dutch workers.[8] Diffuse bands on Laue photographs were at first interpreted as being due to static imperfections in transition regions of a tin crystal, which caused these regions to scatter as a one-dimensional grating. Further investigations, however, showed that the diffuse bands were due to dynamic (thermal) distortions, similar to those from single crystals of aluminium observed by G. D. Preston.[9] Calculations, based on the elastic constants of tin, showed good agreement with observation. (For a further review of this type of imperfection due to thermal vibration, *see* p. 311.)

Interstitial Alloys and Compounds of Metals with Hydrogen, Carbon, and Nitrogen

The detection of the atomic sites of these elements by present X-ray methods requires careful investigation under the best conditions. In cases where the atomic positions cannot be determined directly, some information can be gained by studying the variation of lattice parameters with composition (hydrides), and of other factors which may give indirect evidence of the positions of the interstitial atoms (*e.g.*, martensite).

Hydrogen—X-ray studies of the effect of hydrogen on the palladium lattice have been made by both French[10] and British[11] workers. The French workers observed three stages in the absorption of hydrogen : (1) Hydrogen penetrated the face-centred cubic α phase, and at the same time a phase, β, with the same structure but different lattice parameter, appeared on the surface but did not develop ; (2) after the α phase was saturated, the β phase developed in depth ; (3) when all the metal had become β phase, the parameter of this phase increased. The British

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Apart from the brief reference to neutron diffraction, no material has been added to the review since it was originally written in March, 1947.

workers investigated the hysteresis effect in the formation and decomposition of the β phase for varying temperatures under constant pressure of hydrogen. The effect of varying temperature on the α lattice parameter, the pressure at which the β phase was formed, and other general relationships, were described in detail. It was suggested that the fundamental factor governing the phase change was the concentration of hydrogen in the lattice, not the external hydrogen pressure. No evidence was found for the existence of other phases. It was found that hydrogen could be removed from the palladium lattice at room temperature without reversion of the lattice to its normal state. French workers also studied the system Pd–Au–H.[12] Pd–Au formed only one phase with hydrogen, since the temperature at which the discontinuity between α and β disappeared was then below room temperature; also the metastable structures were much more easily destroyed than in the case of pure Pd.

The effect of hydrogen on the lattice parameter and structure of electrolytic manganese has been investigated, and a linear increase of parameter with hydrogen content was found.[13]

Magnetic and X-ray studies of titanium and zirconium containing hydrogen have been made.[14] Zirconium, with approximately two hydrogen atoms per unit cell, has a tetragonal face-centred structure, $a = 4{\cdot}87$ Å., $c = 4{\cdot}58$ Å. (*cf.* normal structure, hexagonal close-packed with $a = 3{\cdot}23$ Å., $c = 5{\cdot}14$ Å.), and titanium with 1·2 hydrogen atoms per unit cell has an approximately face-centred cubic structure with $a = 4{\cdot}45$ Å. (*cf.* hexagonal close-packed, $a = 2{\cdot}92$ Å., $c = 4{\cdot}67$ Å).

Carbon—Direct evidence has been found for the position of interstitial carbon atoms in the face-centred modification of iron; they are found to occupy the octahedral positions in a random manner in the austenite structure.[15] In the case of martensite, which has a metastable tetragonal structure, evidence has been found for the displacement of iron atoms from the lattice points in the direction of the tetragonal axis, which supports previous ideas as to the positions of the carbon atoms.[16]

By the application of the method of Fourier analysis the atomic positions of carbon in cementite Fe_3C (orthorhombic) have been found.[17] Russian workers have reported the existence of an iron carbide with slightly different cell dimensions and atomic positions from those of cementite.[18] They have also found another carbide, of unknown lattice and composition, formed by tempering martensite below 300° C.

A new method for preparing a carbide of approximate composition Fe_2C has been found; an investigation of the structure of this is proceeding at the Cavendish Laboratory, Cambridge.

The carbides of the iron–manganese system have been studied, and a new monoclinic carbide has been found which does not occur in the Mn–C or Fe–C systems.[19] Two carbides, $Mn_{23}C_6$ and Mn_7C_3, with structures analagous to $Cr_{23}C_6$ and Cr_7C_3, were found; with more than 50% of iron, cementite is the only carbide formed, and manganese can replace iron in cementite to a considerable extent.

Two double carbides have been found in the tungsten–cobalt–carbon systems, both having 96 metallic atoms per unit cell. The compositions are W_4Co_2C and W_6Co_6C, and the positions of the metallic atoms have been determined.[20]

In an investigation of titanium carbides, and titanium–tungsten carbides, it was found that titanium carbide prepared in many different ways was always deficient in carbon, and that there was sufficient oxygen present to fill the vacant lattice positions.[21] No evidence of double carbides with new structures was found.

Nitrogen—By means of accurate intensity measurements and observations of X-ray superlattice lines, the ordered positions of the nitrogen atoms in the interstices of $\gamma'(Fe_4N)$ and $\zeta(Fe_2N)$ phases of the iron–nitrogen system have been determined. The structure of $\varepsilon(Fe_3N)$ is nearly completed, and it is hoped to carry out further work on the γ phase and the nitrides of manganese at the Cavendish Laboratory.

Nickel nitride, Ni_3N, has been prepared and was found to have a hexagonal close-packed arrangement of nickel atoms, with nitrogen in the interstices.[22] Vanadium nitride, VN, was found to be cubic.[23]

Carbo-Nitrides—It has been found possible to replace nitrogen by carbon in the Fe_2N and Fe_3N phases to quite considerable extents; manganese carbo-nitrides have also been prepared. This class of compound will be studied in more detail at the Cavendish Laboratory.

X-ray examination of electrolytically extracted residues from titanium steels showed a continuous series of solid solutions between TiN and TiC.[24]

Intermetallic Phases

Many of the alloys investigated have been found to have simple type structures which are already known. These have been studied in detail by Hume-Rothery, Hägg, and others, and interpreted in terms of such factors as atomic size ratio, electron/atom concentrations, and valency; in particular a large number of the AB_2 type or Laves phases have been investigated. In addition, some alloys with new and more complicated structures have been successfully analysed.

AB_2 Compounds or Laves Phases—In a general article on these compounds, Schulze[25,26] discussed, on the basis of energy and geometrical considerations, what metals were likely to form such phases, and he compared his results with observed data; the high co-ordination number was considered to be a deciding factor for stability. There are three structure types among the AB_2 phases, namely (1) face-centred cubic with eight 'molecules' per unit cell, typified by $MgCu_2$, (2) hexagonal with four 'molecules' per unit cell, typified by $MgZn_2$, and (3) hexagonal, also with four 'molecules' per unit cell, typified by $MgNi_2$. They have no detectable range of homogeneity. Wallbaum[27] has collected the data for such structures formed by the iron, or B, type metals with the A type metals titanium, zirconium, tantalum, and niobium. Alloys of the elements lanthanum, cerium, and praesodymium (A) with copper, silver, gold, magnesium, aluminium, thallium, tin, and lead, were all found to give Laves phases.[28] Thermal,

micrographic, and X-ray studies of these systems now indicate complete analogy between all the intermetallic compounds formed in these systems. Other Laves-type compounds are $CeCo_2$, $CeNi_2$,[29] $CaGa_2$, $LaGa_2$, $CeGa_2$,[30] and $SrMg_2$, $BaMg_2$, $CaLi_2$ ($MgZn_2$ type).[31]

In alloys containing transition elements, there is a tendency for the formation of phases of composition AB_3, although the atomic size factor is suitable for an AB_2 type of compound. Such compounds are Pd_3Ti, Ni_3Ti, Pt_3Zr, all hexagonal close-packed, Pt_3Ti face-centred cubic, and Cu_3Ti, Au_3Ti, Ni_3Nb, Ni_3Ta, all deformed hexagonal close-packed.[32]

Other Alloy Structures—Fe_5Si_3 has been found to have a structure analogous to Mn_5Si_3.[33] From single-crystal photographs the structures of $CuMg_2$, orthorhombic,[34] Al_2CuMg, face-centred orthorhombic,[35] and $Al_8Si_6Mg_3Fe$, hexagonal,[36] have been completely solved.

In a study of the gold–aluminium system by the powder method,[37] six intermediate phases were found; only the structure of Au_4Al could be solved. This has a structure similar to β-manganese. In the iron–tin system, $FeSn_2$ has been found to be tetragonal and isomorphous with $CoSn_2$ and $MnSn_2$, and Fe_3Sn is hexagonal and isomorphous with Ni_3Sn.[38]

The intermetallic compounds previously reported as $CaCu_4$ and $CaZn_4$ were found, in fact, to be $CaCu_5$ and $CaZn_5$, with similar hexagonal structures.[39] A later investigation confirmed these structures and also found the isomorphous compounds $CeNi_5$, $LaNi_5$, $CaNi_5$, $LaCu_5$, $LaZn_5$; at the same time MgCe, MgLa, and MgSr were found to have the CsCl type of structure.[40] $GdNi_5$ was also found to be isomorphous with $CaCu_5$, etc.[41]

In an investigation of manganese–arsenic alloys, it was found that there was a slight atomic rearrangement at the Curie point.[42] In the iron–nickel–aluminium system it was found that an ordered body-centred cubic phase α' was precipitated from α, not a face-centred cubic phase γ, as previously reported; γ appeared to be metastable in this region.[43]

The copper–indium and nickel–indium systems both contain phases with the NiAs type of structure, which can take up excess of indium. This tendency in structures of this type, together with the possibility of the excess atoms being accommodated in the tetrahedral interstices of the structure, has been discussed.[44] The alloys Fe_2Ge, Co_2Ge, Ni_2Ge, Ni_2In, and Cu_2In, also have the NiAs structure. X-ray intensity measurements have shown that the excess atoms are present in the interstices.[45] With a very high proportion of excess atoms, the structure becomes pseudo-cubic and is related to the γ phases Cu_9Ga_4 and Cu_9Al_4.

The β phase in aluminium–magnesium alloys has been found to be face-centred cubic, and not hexagonal, as previously supposed, with 1172 atoms per unit cell, the largest number yet reported for an alloy.[46] Other structures which have not yet been completely solved are Co_2Al_9, which has been found to be monoclinic,[47] and $Al_{12}Cu_{14}$, which has a tetragonal unit cell.[48]

Alloys of nickel and tellurium, with compositions from NiTe to $NiTe_2$, have been studied by the X-ray powder method, and the changes in atomic positions over this range have been discussed.[49] In the copper–zinc system the hexagonal phase has been shown to exist over a composition range which does not fit in with the Hume-Rothery rules.[50] $BiPt_2$ has been found to have the pyrites type of structure,[51] Sn_2Pt has the fluorite type, and VAl_3 has a tetragonal structure resembling $TiAl_3$.[52] Ni_4Mo has a cubic close-packed structure above 900° C., and, on quenching, ordering takes place, resulting in a tetragonal distortion of the lattice.[53]

In the manganese–tin system, $Mn_{11}Sn_3$ is hexagonal close-packed, Mn_2Sn has the NiAs type of structure, and $MnSn_2$ is isomorphous with $FeSn_2$ and $CoSn_2$.[54] Examples of the fluorspar type of structure are CuCdSb and $NiIn_2$, though the latter is tetragonal. Ni_2MgSb has a structure more like the CaF_2 type than has NiMgSb, the excess nickel going into interstices in the structure.[55] More examples of the zinc-blende type of structure are InP, InAs, and InSb.[56]

In alloys of aluminium, Al_4La and Al_4Ce both have a body-centred tetragonal lattice,[57] and Al_7Cr is found to be monoclinic.[58] In a study of magnesium amalgams, six phases have been identified: βMg_3Hg hexagonal, γMg_5Hg_2, δMg_2Hg, and εMg_5Hg_3 hexagonal, ζMgHg body-centred cubic, and $\eta MgHg_2$ body-centred tetragonal with the $MoSi_2$ type of structure.[59] Cu_3Sb has been shown to have a face-centred cubic structure of the Cu_2AlMn type, not body-centred cubic as previously supposed,[60] and LiPb has the β-brass structure.[61]

In the nickel–antimony system, the lattice constants and the number of molecules per unit cell were determined for six phases, all having cubic, tetragonal, or hexagonal structures.[62] All phases present in alloys of titanium with manganese, iron, cobalt, nickel, copper, zinc, gallium, silver, and gold have been investigated in detail.[63]

Defect structures, that is structures in which not all the lattice points are occupied, have been studied in the case of NiAl and of alloys of copper with nickel and aluminium.[64]

LATTICE PARAMETER DETERMINATION

In this section investigations which are concerned solely with determination of the variation of parameters with temperature and composition are discussed. The main interest underlying the latter type of investigation may be one of several things: (1) Measurement of solid solubility (if a standard curve of parameter plotted against composition is obtained it may be used as a method of analysis if the variation in parameter is sufficiently large); (2) interpretation of results with large numbers of different solvents and solutes in terms of such factors as atomic radius-ratio, valencies of solvent and solute, electron concentration, and overlap of Brillouin zones, etc. Lattice parameter measurements are, of course, made in the course of structure or phase-diagram determination, but these are considered separately in the preceding and following sections.

Some improvements in technique have been developed. Improved designs of back-reflection

camera have been published (Guinier, Barraud), and the Straumanis method of placing the film gives accurate results for both single-crystal and powder photographs.[65] Methods of obtaining accurate parameters from single-crystal[66] and powder[67] photographs have been discussed, and, in connection with the latter, new extrapolation functions of the Bragg angle are proposed[68, 69] for graphical elimination of the experimental errors.

Low- and high-temperature cameras necessary for determination of the thermal expansion of the lattice are considered later (p. 311).

Variation with Temperature

The thermal expansions of lead,[70] aluminium,[71] selenium and tellurium,[65] aluminium and gold,[72] and copper,[73] have been measured using high-temperature powder cameras. In the case of aluminium and gold, the thermal expansion was measured by an interferometric method at the same time, complete agreement being obtained between the two sets of results. In the case of copper, good agreement was obtained between X-ray and ' macro ' methods up to 500° C., but above this the lattice expansion was found to be smaller than the ' macro ' expansion ; a possible way of accounting for this is discussed.[73]

The lattice parameter of lithium at 20° C. and at −183° C. has been redetermined using improved extrapolation methods to eliminate experimental errors.[74] Measurements have been made on rubidium at the same temperatures, and the linear coefficient of lattice expansion over part of the range has been obtained.[75] The thermal expansion of $AuCu_3$ has been studied by means of a high-temperature camera at the University College of North Wales, Bangor.[76] The coefficient of thermal expansion of the ordered structure increases rapidly as the critical temperature 385° C. is approached, and then falls discontinuously at the transformation point, where a sudden break in the parameter/temperature curve occurs. No hysteresis phenomenon is observed in passing through this point if the temperature is changed sufficiently slowly.

Variation with Composition

The solid solubility of magnesium[77] and beryllium[78] in aluminium has been investigated by means of parameter measurements combined with electrical resistivity measurements and micrographic examination. In an investigation of sodium in solid solution in lead,[79] it was found that the parameter decreased with increasing concentration of sodium, despite the fact that the atomic radius of sodium is greater than that of lead.

In a discussion of the results obtained when aluminium and indium are dissolved in magnesium, it was concluded that the interatomic distances in alloys cannot be accounted for by assigning a fixed atomic radius to each type of atom, but that overlap of Brillouin zones of electrons must also be taken into account. If there are sufficient electrons to cause an overlap across the side of a zone; the effect is to compress the zone.[80] A qualitative interpretation in terms of the Brillouin zone theory of metals can also be made for lead, bismuth, and gold dissolved in magnesium.[81] At equi-atomic concentrations, contraction in the basal plane increases with decreasing valency of the solute. Both the a and the volume parameters can be expressed as functions of the valency of the solute. In an extension of the work on aluminium and indium to the other Group III metals, and also tin, dissolved in magnesium,[82] the effect of increasing valency on the increasing volume of the lattice was confirmed. The c-spacing changed discontinuously at an electron concentration at which the first Brillouin zones overlapped in the c-direction (the zones already overlap in the a-direction in pure magnesium). For solutes of favourable size factor, the variation of axial ratio with composition was proportional to the electron concentration after the onset of the overlap in the c-direction. The apparent atomic radii, as deduced by extrapolation of lattice parameter composition curves to 100% solute, were in agreement with the requirements of the Brillouin zone theory. Results confirmed the suggestion that indium, thallium, white tin, and lead, were incompletely ionized in the elementary state.

Some work has also been done in America on the solubility of lead, bismuth, and gold in magnesium.[83] In an investigation of calcium–strontium alloys, it was found that the lattice parameter varied linearly with composition across the whole continuous range of solid solutions.[84] A comparison of lattice parameters and densities of α phase zinc–silver alloys indicated that at about 3, 17, and 29 at.-% of zinc, the lattice sites were all occupied, but at intermediate compositions a very small number were unoccupied.[85]

Measurements of the α-iron lattice parameter have been made for known concentrations of molybdenum in order to establish data for the estimation of molybdenum, and hence to throw light on the problem of the partition of molybdenum in steel. This problem has also been the subject of two later investigations.[86] Similarly accurate data have been established for the α phase of iron–silicon alloys.[87]

In alloys of copper or silver with many of the B sub-group metals, additions of solute result first in a solid solution α, then a two-phase region $\alpha + \beta'$ (ordered) or $\alpha + \beta$ (disordered). The α/β type of equilibrium diagram is quite different from the α/β'. Equilibrium between α and β is known to depend largely on electron concentration, but atomic size-factors and the resultant lattice distortions have some effect. Some general conclusions on the latter were drawn from high-temperature investigations of two-phase alloys in the systems Cu–Zn, Cu–In, Cu–Sn, and Cu–Ga.[88] Alloys of the first and second long-period elements in copper, silver, and gold, have also been investigated in detail at Bangor.[89] Some interesting relations between several factors such as average percentage distortion and valency of solute, distortion and nature of solvent, distortion at maximum solubility and total volume of solute entering the lattice, electron concentration at maximum solubility and valency of solute, and maximum solubility and atomic radius-ratio, were noted. Where the valency of the solute atom was three or less, the atom size controlled the distortion of the solvent lattice, but for higher valencies the distortion was controlled by the valency.

PHASE DIAGRAMS

Several articles have been written, and discussions have been held, on the use of X-rays in the determination of phase diagrams.[90–92] The X-ray method has some advantages such as the possibilities of examining specimens at high temperatures, and of providing an accurate identification of phases as well as the data from which the atomic structure (including ordering) of the phases can be deduced. It also has several limitations : it is not capable of detecting very small quantities of a second phase, except under very special conditions, and it cannot be used for determining the solidus and liquidus boundaries. It reveals little about the distribution of the phases, and furthermore the work is usually performed on powder specimens which may not behave in the same manner as massive specimens. The general conclusions are that the best results for phase-diagram analysis can be obtained by a suitable combination of X-ray and other methods.

The two X-ray methods of determining phase boundaries, parametric and disappearing-phase methods, are discussed in Chapter X of C. S. Barrett's "*The Structure of Metals*" (1943). In the following investigations, some workers used X-ray methods only (references in bold type) and some used X-rays in conjunction with other methods.

Binary Diagrams

Iron, Nickel, and Cobalt Alloys—The Fe–Si diagram has been investigated by Japanese[93] and British workers.[87] The latter found a new body-centred cubic phase α'', stable between 1030° and 1195° C., which caused them to modify the phase diagram. The reaction by which α'' is formed is unusual, and by consideration of free-energy principles it was concluded that the reaction was either polymorphic or peritectic, but not peritectoid. An interesting relationship was found between the orientation of the η phase and that of the α matrix ; on this basis a reason was advanced for the precipitation of η outside its range of stability.

In the course of a determination of the phase boundaries in Fe–Co,[94] an order–disorder change was found in the alloy FeCo with $T_c = 700°$ C. The variation of the lattice parameter of the α phase deviated considerably from a linear function of cobalt content. The Fe–Sn,[95] Fe–Pd, Ni–Pd,[96] and Fe–Mn[97] systems have been investigated by American workers. In the last case, the $\alpha/\alpha + \gamma$ phase boundary was determined entirely by the X-ray parametric method ; it was found that α was apt to occur in the supersaturated state, and that ε was a metastable state. Investigations of the Ni–Mn and Cu–Mn systems[98] were concerned mainly with measurement of physical properties. The alloy MnNi becomes brittle, and its resistivity drops, on ageing at 500° C. ; X-ray and micrographic evidence attributes this both to an order–disorder change and to the precipitation of a second phase. The corresponding drop in resistivity for an alloy CuMn is associated only with a contraction of the lattice (disorder–order) ; no evidence was found for a $CuMn_2$ superlattice. Later investigations of the Cu–Mn system[99] resulted in a phase diagram differing from previous results in the positions of several of the boundaries.

The Fe–Ni system has been investigated by X-rays[100,101] ; it is very complicated despite the fact that only body-centred and face-centred cubic structures are present. Further work, using X-ray methods only, has been done on both high-temperature and quenched specimens at Bangor, confirming previous results apart from a slight alteration of the $\alpha + \gamma/\gamma$ boundary at high temperatures.

The systems Fe–Ti, Ni–Ti, and Co–Ti have been completely re-examined by thermal and X-ray analysis. Portions of the systems of Fe, Ni, Co and Mn with Ti, Zr, Nb, and Ta, were studied by X-rays in order to determine the composition of the intermediate phases.[102]

In the system Co–Zn, the $\gamma/\gamma + \beta$ and $\gamma/\gamma + \beta'$ phase boundaries were determined by X-ray and micrographic analysis, and lattice parameter measurements were made for the range of solid solutions of zinc in face-centred cubic cobalt.[103] The system Al–Co was investigated by thermal, micrographic, X-ray, and magnetic methods.[104] The system Pt–Co has also been studied in some detail,[105] together with Pt–Mo and Pt–W.[106] The alloy PtCo was found to have a tetragonal superlattice structure at 600° C. (*cf.* PtFe*), becoming disordered face-centred cubic at 1000° C. and body-centred cubic at 400° C.

In a study of the Ni–Sn system, three intermediate phases were found, of which Ni_3Sn_2 had the NiAs type of structure and Ni_3Sn could exist in the ordered state below 900° C.[107] No evidence was found for a phase Ni_4Sn. The systems Ni–Mo[108] and Ni–Al[109] have been investigated by means of several other methods in conjunction with X-rays, as has also the Ni–W system.[110]

Copper, Silver, and Gold Alloys—Work has been done on the Cu–Si system by American[111] and Japanese workers.[112] The latter point out that the α, β, and γ phases are analogous to phases occurring in other binary copper systems, and they conform to Hume-Rothery's rules, while κ and δ are peculiar to this system. Both give diagrams for up to $\sim$ 14 at.-% of silicon. The Cu–Pd phase diagram was determined by X-rays in conjunction with previous resistivity measurements[113] ; an order–disorder transformation was found in Cu_3Pd. Alloys of Cu–Sb from 60 to 70 wt.-% of copper were studied by X-rays and with the microscope, and the results confirmed earlier diagrams[114] ; the ε phase had a hexagonal close-packed structure, and $\beta Cu_{32}Sb_4$ was cubic, possessing an ordered structure resembling $FeAl_3$.

The Ag–Cd system has been determined entirely by the X-ray method,[115] and all the phase structures have been solved. In a study of the Ag–Zn system,[116] results were compared with those of other workers and also with results for Ag–Cd. The application of X-ray methods to this type of alloy system was discussed, and their limitations were indicated. A preliminary study of the α and β solid solutions has been made in the Ag–Mg systems,[117] and in the system Ag–Pt[118] a phase Ag_3Pt has been found which can exist in the ordered state below 800° C. ; there appeared to be two phases of composition AgPt.

The α phase boundaries have been determined for a

* *Journal of the Institute of Metals*, 1941, vol. 67, p. 333.

large number of binary systems with gold,[119] Au–Ge[120] being studied particularly in the gold-rich portion. The Au–Cd system has been investigated by several workers.[121, 122] The latest work by Dr. Bystrom and his collaborators (*Acta Chemica Scandinavica*, in the press) gave results differing from previous work in several respects : (*a*) The α' phase had a tetragonal ordered structure and transformed to a cubic ordered structure α'' at 430° C. ; and (*b*) the α_3 phase was not found. Instead, in this region, an orthorhombic phase β' of known structure existed at low temperatures, and a cubic β phase with CsCl structure existed at higher temperatures. The α_2 phase had an ideal hexagonal close-packed structure only at the composition Au_2Cd ; at higher gold contents there were occasional faults in the structure, probably of the type occurring in cobalt.

Miscellaneous Binary Systems—A detailed X-ray study has been made of alloys of tin with transition metals,[123] though with emphasis on structure rather than on phase-boundary determination ; numerous corrections have, however, been made in the previous diagrams. Several different methods have been used in researches on the Mn–Zn system,[124] and, in a study of the Ca–Sr, Ca–Ba, and Sr–Ba systems,[125] it was found that calcium and strontium formed a continuous series of solid solutions, but two-phase regions occurred in the other two systems. X-ray methods only were used on the system Al–Sr,[126] and two intermediate phases, Al_4Sr, body-centred tetragonal, and AlSr, body-centred cubic, were found.

In the system Mg–Cd, the presence of $MgCd_2$ was not confirmed[127] ; three phases were found in the Al–Mg system.[128] Pt–Cr[129] was studied by means of the microscope and X-rays. A miscibility gap was detected in the In–Ti system,[130] and the lattice types of all phases present were determined. A tentative constitutional diagram based on metallographic and X-ray observations has been given for part of the Ta–C system, including two carbides, Ta_2C and TaC.[131]

Ternary Systems

A discussion of ternary diagrams and the use of X-rays in plotting these is given in a general discussion of researches into the structure of alloys.[132] The ways in which this method can be used to minimize the large amount of work involved are outlined, and a summary of results obtained up to the year 1940 is given. It should be noted that in several of the investigations mentioned below, the binary diagrams were investigated as a preliminary to the work on the ternary system.

Aluminium Alloys—Many investigations have been made on ternaries containing aluminium ; the aluminium-rich corner of the diagram tends to contain a large number of complicated structures, many of which have not yet been solved. At Birmingham University, G. V. Raynor, developing and extending researches initiated in Oxford, is studying the transitional metals of the first long period in alloy formation with aluminium ; examination of the aluminium-rich portion of several systems has shown that both atomic and electronic factors are operative in determining the constitution, and these are being more closely examined. Detailed investigations have already been made of Al–Mn–Zn, Al–Fe–Ni, Al–Ni–Co, and Al–Mn–Co, and a study of Al–Mn–Cd, Al–Mn–Mg, Al–Fe–Co, and Al–Co–Cu, is in progress. In this work, X-ray methods are used in conjunction with the microscope and other methods.

Aluminium-rich Al–Cr–Mn alloys have been studied by micrographic and X-ray methods[133] ; for the Al–Mg–Cr system, only the X-ray method was used,[134] as was also the case in the Al–Mg–Si system.[135] The Al–Fe–Mn system has been studied by micrographic, thermal, and X-ray methods.[136] Alloys of Al–Co–Fe with less than 50 at.-% of aluminium were examined by means of X-rays, in the state as quenched from 800° C. ; results could not be fully explained and further work is necessary.[137]

In the system Al–Mg–Cu, full use was made of the X-ray method, supplemented by the microscope.[138] Later investigations of this system, together with Al–Cu–Mn, Al–Sb–Sn, and also Ag–Cu–Mg and Cu–Ni–Co, were made using X-ray methods only.[139]

Iron, Nickel, and Cobalt Alloys—X-ray methods have been used in conjunction with other methods on iron ternary diagrams by German workers ; among the systems they have studied are Fe–B–S,[140] Fe–Al–S,[141] Fe–Ti–Cr,[142] and Fe–Si–Va.[143]

Fe–Ni–Cr alloys were studied by means of X-rays after slow cooling.[144] It was found that the narrow range of composition of the σ phase in the Fe–Cr system was greatly increased by the addition of nickel. In the Fe–Cu–Ni system[145] the X-ray parametric method was used to determine the α phase boundary and $\alpha + \alpha'$ tie-lines, in contrast with the method of estimation of relative quantities of phase used in the investigation of Fe–Ni–Cr. The metastable states found in one region of the $\alpha + \alpha'$ phase field led to the study of the dissociation of the α phase (*see* p. 307).

In the Fe–Ni–Si system, two new phases were found and their structures were tentatively identified.[146] The system Fe–Mn–C has been studied by X-rays only, up to 50% of iron and 1·5% of carbon ; the ε phase (hexagonal) appeared to be metastable, since strong deformation caused it to disappear.[147] This was similar to results obtained with Fe–Mn.[97]

Ni–Co–Al[148] and Ni–Cr–C[149] systems were both studied by several methods, including X-rays ; in the latter no ternary compound was found.

Copper, Silver, and Gold Alloys—At Birmingham, work is in progress on the mutual effect of various theoretical factors on the constitution of ternary copper- and silver-rich alloys ; the high-temperature X-ray powder camera is being used in this work.

X-ray methods were used exclusively on the Cu–Fe–Si system,[150] and in contradiction to earlier results it was found that the lattice of copper was expanded by iron in solid solution ; results for the system were very complex. Japanese workers have studied the Cu–Sb–Ni,[151] Cu–Al–Fe,[152] Cu–Ni–Be,[153] and Cu–Ni–Si[154] systems, using X-rays in conjunction with other methods. X-ray investigations have been made of the sections Ag_5Sn–$AgCd_3$ and Ag_5Sn–AgCd in the Ag–Cd–Sn system.[155]

Quaternary Systems, etc.

Work is being done at Oxford on the system Al–Mg–Mn–Zn and on the effects of silicon, copper, and chromium on this type of alloy. Some ternary and quaternary systems have been investigated as a first step towards the elucidation of the quinary systems. The system Zn–Mn–Cu–Ag has been investigated by means of X-rays and the microscope; it was found that the ε phases of Zn–Mn, Zn–Cu, and Zn–Ag, formed a continuous series of solid solutions with each other.[156]

TRANSFORMATIONS IN ALLOYS

X-Ray Investigation of Age-Hardening

The phenomenon of age-hardening occurs in binary alloys where the solid solubility of one metal in the other increases with temperature. If an alloy, which is two-phase in equilibrium at room temperature, is heat-treated so that it becomes single-phase, and is then quenched, the precipitation of the second phase occurs slowly (the rate depending on temperature of ageing) and can be followed in detail by the X-ray method. The gradual phase change is accompanied by changes in physical properties, such as increase in hardness and in electric resistance, which reach maximum values at certain stages of the reaction. Examination of the microstructure shows that Widmanstätten structures tend to develop, though in some cases no change in microstructure is detected even when maximum hardness is attained. This can sometimes be remedied by the use of extremely careful metallographic technique.

The X-ray method is capable of detecting the size and shape of the first-formed regions of concentration of solute atoms, and also the structure, and orientation relative to the matrix, of any phases which may later develop in these regions. The lattice parameter of the matrix must be studied with care, since results will depend upon the fineness of distribution of regions of precipitation, though initial increases in lattice parameter have been claimed to show ordering throughout the matrix.[157]

The several successive stages in the process of lattice alteration which converts the matrix lattice to that of the equilibrium phase, can be divided up as follows:

(1) The solute content increases in small regions parallel to certain matrix planes.

(2) Thin plate-like regions are produced which form two-dimensional diffraction gratings. These may be regarded as first consisting of superlattice structures which later change over, by simple atomic movements, to a transition structure which is coherent with the matrix across the habit plane.

(3) These particles, with transition structures, grow in thickness and in lateral dimensions, in some alloys becoming only a few atoms thick, and in others attaining a size which produces three-dimensional diffraction and enables the actual structure to be determined.

(4) On reaching a certain size the transition structure transforms to the stable precipitate structure.

Hardening occurs during the first stages and reaches a maximum at the end of stage (3), and it is now considered that the strains accompanying the existence of the transitional structure (*i.e.*, maintaining coherence across the matrix-precipitate interface) are the major factor in the phenomenon of age-hardening. This is in accord with the observation that Al–Cu, Al–Ag, and Al–Mg, which form coherent transition lattice regions of considerable size, age-harden appreciably, whereas Al–Zn, Mg–Pb, and Mg–Sn, in which the coherent particles break away from the matrix when very small, do not harden appreciably. The extent of coherent growth appears to be related, as would be expected, to the amount of atomic movement involved in the change from transition to equilibrium structure. This is shown by the difference in size of transition regions in Al–Ag and Al–Zn, where the habit plane and equilibrium lattice type are the same, but the disregistry for Al–Ag is 0·7% and for Al–Zn 7%. The problem of extent of coherent growth is complicated if the transition lattice differs greatly from the equilibrium lattice as in Cu–Al. The number, as well as the size, of the coherent particles, affects the degree of hardening, and hence the hardening will vary with temperature of ageing.

Aluminium–Copper—X-ray work was done on this system before the war[158] and was later extended.[159] The transition-lattice θ′ regions have a slightly distorted CaF_2 type of structure, and form on the (100) matrix planes with [100]θ′ parallel to [110] matrix. The equilibrium precipitate $CuAl_2$ is tetragonal, and has a multiplicity of orientations relative to θ′, and it has not proved possible to derive a probable atomic mechanism for the transition.[160]

An X-ray study of the time-rate of precipitation has been made, using a Geiger counter to follow the changes in intensity of the diffraction lines. Results support the view that the transition phase changes rapidly at 309° C. to the 'randomly' oriented final phase, which finally exists in equilibrium with a small amount of the transition phase.[161]

Aluminium–Silver—In this case plate-like regions form on the (111) matrix planes, and the transition and final lattices are both hexagonal, approximately close-packed with disordered arrangements of atoms. Ordering has been detected in the first-formed small regions.[159,160] The atomic movements that transform the matrix to the final precipitate have been inferred from the orientation relationships involved, and the stresses set up by the strained lattice have been computed.[162]

Aluminium–Magnesium—Difficulty was experienced at first in detecting the streaks on Laue photographs which show the existence of precipitation platelets, but these have been found.[163] The platelets appear to occur on both (100) and (120) matrix planes, and grow to sufficient size to produce a three-dimensional grating.[164] The structures of both transitional and final phases are complicated and are not known with certainty.

Aluminium–Zinc—This resembles Al–Ag, except that the final precipitate is practically pure zinc.[163,165] The transitional regions do not grow large enough for direct evidence as to the structure to be obtained. French workers have used a micro-radiographic technique to show that the platelets consist of pure zinc.

Copper–Beryllium—This resembles Al–Cu in that the first regions of concentration form on the (100) matrix planes. The plates do not attain large diameters and break away before the thickness is such that they act as three-dimensional gratings. The precipitate formed is very fine-grained and the orientation is more or less random. No Widmanstätten structures are observed under the microscope.[166]

Magnesium Alloys—Mg–Zn, Mg–Pb, and Mg–Sn were studied in order to find whether the coherent platelets grow to sufficient size to give a three-dimensional diffraction pattern.[163] No platelet streaks were observed for Mg–Pb though there was an increase of blackening round the central portion of the pattern, possibly indicating small, roughly spherical, particles. With Mg–Pb and Mg–Sn, the coherent platelets break away when small, and the first observed diffraction pattern is that of the equilibrium precipitate ; this agrees with the small age-hardening observed. The platelets in Mg–Sn form parallel to three types of high-order matrix planes with high disregistry for all three types, and the large number of possible orientations for the final precipitate results in almost random orientation.

In Mg–Zn, where the age-hardening is considerable, the diffraction pattern of the transition structure was unfortunately not observed as would be expected (*cf.* Cu–Be).

Copper–Gold—This differs from the above systems in that the age-hardening is not due to the precipitation of a second phase, but to an order–disorder reaction ; the ordered structure is tetragonal and the disordered structure is cubic. The ordering of a small region gives rise to internal stresses such that further ordering would be impossible unless other regions had orientations differing from that of the original. For minimum total strain, these regions are arranged so that, in a layer parallel to a (111) matrix plane, all regions have their tetragonal axes parallel to one original cube edge ; in the next layer the tetragonal axes are parallel to a different original cube edge, and so on. X-ray results confirm this arrangement, and the structure is in a state of internal strain which is correlated with its hardness.[167]

Iron–Tungsten—A search for evidence of the formation of platelets gave negative results.[168] The three-dimensional diffraction pattern of the precipitated phase was such that it was not possible to detect whether it was a transition phase, or either of the equilibrium phases Fe_7W_6 or Fe_2W. Knowing the structures of these two phases it was possible to deduce some information about the atomic movements involved in the initial formation and perfecting of the structure, which was confirmed by the relative broadening of the diffraction lines.

In all the above investigations the main interest had centred on the transition and equilibrium forms of the precipitate. In earlier work, measurements were made on the lattice parameter of the matrix, and transformations were listed as continuous or discontinuous according to whether the parameter appeared to change gradually or discontinuously. In the case of iron–tungsten alloys, the change appears to be discontinuous, and lines due to the depleted matrix appear as hardening starts, and those of the supersaturated matrix disappear as hardening attains its highest value.

Magnesium–Aluminium–Bismuth—In this case, also, the change is followed by observing the changes in matrix lattice parameter, which are continuous.[169] The formation of Mg_3Al_2, which is precipitated first, causes an increase in the parameter, and the later formation of Mg_3Bi_2 causes a decrease. The hardness rises as Mg_3Al_2 is precipitated (whether in equilibrium or transition state is not known), attains a constant value, and then rises again as Mg_3Bi_2 appears ; the habit plane of the precipitates is the magnesium basal plane.

Copper-Rich and Silver-Rich Alloys of the Silver–Copper System—In this case the problem is complicated by the fact that the matrix and precipitate have the same structure ; and as the orientations of matrix and precipitate are the same, the misfit would be equally bad on all planes of contact of the matrix and precipitate. It is thus unlikely that platelets would form on any particular set of matrix planes, and this accords with the fact that Widmanstätten patterns are much less readily developed than in Al–Cu and Al–Ag.

No evidence of platelet formation was found,[170] but experiments were performed on single crystals, and also on polycrystalline specimens, in order to study factors influencing the continuity of the precipitation. It was concluded that (1) a crystal in which discontinuous precipitation is taking place is composed largely of regions in which precipitation is complete or has not commenced, with a few regions in the intermediate stage ; (2) discontinuous precipitation occurs when the supersaturation is high and the temperature is low ; (3) fragmentation of the matrix accompanies the discontinuous precipitation ; and (4) it appears that such precipitation takes place only in alloy systems where there is a large difference between atomic sizes of solute and solvent. Thus it appears that discontinuous precipitation is associated with large lattice strains, which is in accordance with the theoretical work of Nabarro,[171] who has discussed the mechanism of discontinuous and continuous precipitation in some detail.

Dissociation in Copper–Iron–Nickel Alloys

The equilibrium diagram for copper, iron, and nickel, shows a large region in which two face-centred cubic phases are in equilibrium. These two phases can be classed as copper-rich and copper-poor, the former having the larger lattice parameter. The tie-lines become shorter as the nickel content of the alloy increases, and there is a critical composition at any one temperature, for which the tie-line is of zero length. In the neighbourhood of this composition the structures of the alloys undergo some interesting changes with heat-treatment, and an investigation of these has provided new information on the process of dissociation in the alloys.

The alloy Cu_4FeNi_3 has been taken as typical.[172] Above about 700° C. it is single-phase (face-centred cubic), and below this temperature it dissociates into the two face-centred cubic phases. The process of dissociation is slow : at 680° C. it takes several months to complete ; it can therefore be followed by

taking X-ray powder photographs of specimens quenched after different annealing periods. After only about an hour certain changes are observed ; at the sides of each powder line there appears a pair of faint diffuse lines, which have been called side-bands. With increase of annealing time these bands become stronger and sharper and move closer in to the main lines. Then the main lines become weaker and other satellite lines appear until, at a certain stage, each powder line is replaced by a group of lines of varying degrees of complexity, the lines with Bragg angle near 90° showing the most marked effects. After this the pattern becomes less complex ; the groups of lines split gradually into the two sets of lines representing the two face-centred cubic phases that form the end-point of the reaction.

The following explanation has been put forward to account for these effects. Since the copper atom is appreciably larger than the other two, the first effect of the migration of these atoms is to produce an enlargement of the lattice at the points where the copper concentration increases, and a contraction at the points that the copper atoms are leaving. This effect is comparable with faulty ruling of an optical diffraction grating ; if the faults in ruling are periodic, then satellite lines are produced. It is suggested that the X-ray side-bands are of this nature, that is, that the copper atoms precipitate out at regular intervals.

Detailed investigation supports this theory. Measurement of the position of the side-bands gives the wavelength of the periodicity, and measurement of their intensities gives the variation in copper concentration. These quantities have been examined as functions of time of heat-treatment at different temperatures. It appears that the difference in copper concentration characteristic of the final two-phase state is reached almost immediately, but that the scale of distribution is very fine. As the annealing progresses, the difference in concentration remains constant, but the scale of distribution increases in size. This process continues from about 100 Å., for which value the side-bands are just detectable, to visible regions—about 5000 Å.

This investigation shows how it is possible to obtain information about diffusion in alloys on a scale above that usually considered possible with X-rays, but below that possible with visible light. The phenomenon, of course, is not likely to occur generally, but it has also been noted in iron–nickel–aluminium alloys, and if it proves to be of frequent occurrence it may provide interesting data about the process of atomic migration during phase changes.

The Mechanism of Phase Change

A considerable amount of work was done in this field before the war, and a table summarizing the results will be found in C. S. Barrett's "*The Structure of Metals.*"* In the foregoing discussion of age-hardening, it was pointed out that investigations on this subject involved the determination of relative orientations and habit planes, and in some cases atomic mechanisms of the phase changes have been suggested (*e.g.*, Al–Ag). This short review covers only some recent work done on steels and brasses.

Ferrous Alloys—Using precision technique, the orientation relationships between austenite (γ face-centred cubic), martensite (body-centred tetragonal), and ferrite (α body-centred cubic), have been determined for a 22% nickel, 0·8% carbon alloy.[173] It was found that the martensite (110) planes deviated from the austenite (111) planes by about 1°, but otherwise the relative orientations were midway between the Kurdjumow–Sachs and Nishiyama relationships. (These two sets of workers had found two slightly different relative orientations, *i.e.*, $(11\bar{1})\gamma//(1\bar{1}0)\alpha$, $[110]\gamma//[111]\alpha$, and $(111)\gamma//(110)\alpha$, $[\bar{2}11]\gamma//[\bar{1}10]\alpha$ respectively, on the basis of which they had proposed two slightly different sets of atomic movements, though both involved a shear on $(111)\gamma$.) Based on micrographic observations of martensite, a mechanism of two shears together with slight dimensional adjustments was suggested, the first on the irrational habit plane of martensite[173] and the second on the (112) martensite planes.

A comprehensive investigation of the orientations of martensite, bainite, and ferrite, relative to austenite in eutectoid steels, was carried out. Bainite formed at 450° and 350° C. displayed the Nishiyama relationship ; bainite formed at 250° C., and also martensite, displayed the Kurdjumow–Sachs relationship.[174] The habit plane changed continuously throughout the temperature range. The orientation of the ferrite in pearlite was not easy to determine, but it appeared that for both fine and coarse pearlite the relative orientation was quite different from that of ferrite in bainite or pure iron, and was either $(110)\gamma//(112)\alpha$, $[112]\gamma//[110]\alpha$ or $(321)\gamma//(331)\alpha$, $[111]\gamma//[321]\alpha$.

Russian workers have studied the relative orientation of cementite and austenite, which they found to be $(010)Fe_3C//(011)\gamma$, $[00\bar{1}]Fe_3C//[5\bar{2}\bar{2}]\lambda$.[175, 176]

A considerable amount of work has been done by R. F. Mehl and his collaborators in America, on the nucleation mechanism of phase-transformation and rate of growth of nuclei, *e.g.*, as in the formation of pearlite, but as this does not lie in the field of X-ray investigations it will not be considered here.

Investigations of the face-centred cubic α phase and body-centred cubic β phase in the iron–nickel–aluminium system showed that the Kurdjumow–Sachs' relationship was obeyed.

Brasses—An ingenious method was devised for forming a layer of β on a single crystal of α-brass, of α on β, β on γ, and γ on β.[177] β on γ and γ on β took up orientations in which all cube axes in the two structures were parallel. α on β and β on α adopted orientations clustering about those found both by Kurdjumow and Sachs and by Nishiyama. For α on β the same relative orientation existed when β was formed from α by a peritectic reaction, or when the phase was formed by precipitation.

Transition Structures in the Copper–Manganese System

Certain physical properties of Cu–Mn alloys were found to be inexplicable on the basis of the Cu–Mn equilibrium diagram. X-ray experiments suggested that the course of the decomposition of the γ solid solution was responsible.[178] The proposed course of

* *Loc. cit.*, p. 486.

the decomposition was complicated, involving several transitional structures, namely $\gamma \rightarrow$ single-phase tetragonal structure of variable axial ratio $\rightarrow$ tetragonal ε phase, axial ratio 0·960, + α Mn; then ε phase $\rightarrow \alpha$ Mn + non-equilibrium cubic phases, and finally non-equilibrium cubic phases $\rightarrow \alpha$ Mn + equilibrium cubic solid solution.

GROWTH, DEFORMATION, AND RECRYSTALLIZATION TEXTURES IN METALS

X-ray methods are being increasingly used for the investigation of preferred orientation in worked metals, a subject which is of great importance in industry because of its profound effect on the variation of physical properties with direction. Results are best expressed in terms of pole-figures for one or more of the atomic planes ; these pole-figures are then interpreted in terms of one or more textures. A typical sheet texture for a body-centred cubic structure is represented by [100] (110), which means that the crystallites tend to be orientated so that the normal to the (100) planes is parallel to the direction of rolling, and the (110) planes are parallel to the plane of the sheet (rolling plane), but with a small angular spread about this orientation. The X-ray method usually involves taking a considerable number of photographs, but recently methods have been developed which enable sufficient information for one pole-figure to be obtained from one photograph.[179,180,181]

According to simple theory, the mechanism of re-orientation during deformation involves a modification of the crystals by slip and by twinning ; the processes are controlled partly by the properties of the crystals, and partly by interaction at their grain boundaries. The first factor had been studied in some detail before the war by investigations, involving use of X-rays, on deformation of single crystals (*e.g.*, Andrade and Elam). Several possible glide planes were found, especially for body-centred cubic metals, which were operative under different conditions, but the glide direction was always that of closest packing.

However, the conditions in polycrystalline sheet, and the number of possible variations of external factors, are very complex, and a complete understanding of the texture developed in terms of the mechanisms of re-orientation is not possible at the present time.

In an attempt to fill this gap, American workers have mounted single crystals of iron[182] and copper[183] in polycrystalline blocks which were then cold-rolled. In the case of α-iron, it was found that the single crystal developed one or more of the following textures: (001) [1$\bar{1}$0], (115) [110], ($\bar{1}$13) [110], ($\bar{1}$12) [110], ($\bar{1}$11) [110], ($\bar{1}$11) [211], and that the final positions of the cube poles, when superposed on the cube pole-figure for the polycrystalline sheet, showed remarkable agreement. In the case of copper, large-grained material was also investigated. It was found that, in general, multiple orientations developed in nearly every crystallite except those with (110) initially parallel to the rolling plane, and [$\bar{1}$12] parallel to the rolling direction ; two crystals retained this orientation after reductions of 97·6%. The multiple orientations developed did not in general correspond with twins, and, though the symmetrical (110) [112] positions were in twin relationship, this was believed to be a coincidence. The principal textures found in rolled polycrystalline material are (234) [32$\bar{3}$] and (135) [2$\bar{1}$1].

Other workers[184] have found a deformation texture of (110) [$\bar{1}$12] and (110) [1$\bar{1}\bar{2}$], or "twin-fibre," texture for rolled copper. After annealing to produce recrystallization, the texture became random if the reduction was less than 50%, or for greater reductions either retained the twin-fibre structure or became single texture (100) [100]. Which of the latter types was present, or relative amounts of the two, depended mainly on the grain size before rolling and the magnitude of reduction, and not on annealing temperature. It has also been found[185] that sheet in which the grain size had been reduced to 0·0008 mm., or less, recrystallized at room temperature with single cubic texture. The twin-fibre structure has also been confirmed by other researches.[186] In this case, experiments were also made on single crystals of copper in which it was found that in adjacent layers the lattice rotated in opposite directions until, after very severe rolling, the texture of polycrystalline copper was approached. It has been reported that the texture of cold-rolled copper strip contained, in addition to (110) [1$\bar{1}\bar{2}$] and (110) [$\bar{1}$12], some crystals in the (100) [100] or recrystallization texture[187,188] (*cf.* work on Ni–Fe sheet, *see* page 310).

The latter workers also investigated α-brass, which they found had the same deformation texture as 70/30 brass, namely, the twin-fibre structure as in copper. The texture in heavily cold-rolled 70/30 brass has been interpreted as (113) [112], but other interpretations are possible.[189] It has been found that the texture developed in lightly rolled brass is (135) [2$\bar{1}$1] (*cf.* Barrett's work on copper). This is also the texture developed in aluminium. These workers confirmed the twin-fibre structure for severely rolled brass.[190] In a detailed investigation of both deformation and recrystallization textures of 70/30 brass, it was concluded that the complex structure of the latter, (113) [112], arose from the former by a process of rotation and of twinning.[191] Twin relationships have also been found in annealed copper strip.[192] In severely cold-rolled strip annealed at low temperatures, the recrystallization texture contained, in addition to the cold-rolled twin-fibre structure and (100) [100], the texture (120) [100] ; annealing for longer periods at higher temperatures produced a texture consisting of (120) [100], plus all first-order twins of this orientation. Work on cartridge brass containing copper, zinc, iron, and lead[193] showed that the cold-rolled sheet had the (110) [$\bar{1}$12] twin-fibre structure, and that, on annealing at low temperatures, the (113) [112] texture was formed ; re-annealing at high temperatures gave the twin-fibre texture again.

Research on aluminium indicated the same (112) fibring as for copper and brass, but the method of cold-work influenced the texture. Cold-rolled aluminium of commercial purity usually recrystallized with random orientation, though the iron content helped to retain the deformation texture. High-purity

aluminium may give a (100) [100] recrystallization texture.[194] Aluminium showed very great plasticity in rapid deformation, but X-rays showed that the texture was not affected by the rate of deformation.[195]

Investigations of magnesium sheet all showed that the basal plane of the hexagonal unit cell tended to lie parallel to the plane of the sheet, giving a (0001) [$11\bar{2}0$] texture. There is, however, an asymmetrical spread about this orientation which has been described in various ways.[196] It was thought that in alloys containing calcium and manganese, a double texture tended to develop, with [0001] lying $\pm 15°$ to the sheet normal.[197] Later work revealed that this double texture was present even in pure magnesium; the surface had the (0001) [$11\bar{2}0$] texture, but the double texture developed with increasing depth below the surface.[198] In a detailed investigation of the twinning taking place in rolled magnesium sheet, it has been shown that re-orientation by twinning on ($10\bar{1}2$) resulted in the hexad axis becoming parallel to the direction of compression. Thus, by combining cold compressions in two or more directions in the plane of the sheet with slow hot-rolling (when only slip, not twinning, takes place), it is possible to produce a sheet without any preferred orientation.[199]

Preferred orientation in Fe–Si alloys had been the subject of several investigations before the war, because of its use as transformer sheet steel. Alloys containing about 3% of silicon, hot-rolled and then subjected to cold reduction in two stages, followed by recrystallization, were found to have the desired (110) [100] texture.[200] The two cold-rollings, combined with two high-temperature ($\sim 1100°$ C.) anneals, appeared to be responsible for this result.[201] Another investigator has reported that in 3·45% silicon sheet, rolled and annealed so that recrystallization takes place, the orientation developed could not be accounted for by a purely random process, but agreed with a second- or third-order twin relationship.[202] The texture of low-carbon steels produced by severe hot-rolling between A_1 and A_3 has been studied, and the results were very complex. Cold-rolling previously hot-rolled strip was found not to affect the texture.

The rolling and recrystallization textures of Ni–Fe sheet with high nickel content have been studied in some detail, together with those of aluminium and copper.[181,203–205] The rolling textures of both aluminium and Ni–Fe would be described as a mixture of two principal orientations, (110) [$\bar{1}12$] and (112) [111]; there was, however, a small difference in that Ni–Fe, after severe cold-rolling, had some crystal grains in the (100) [001] orientation (*cf.* copper). It has not so far proved possible to detect the origin of these crystals, that is, to detect any difference between the deformation processes in the two types of sheet. The recrystallization textures were very different as a result of this slight difference in deformation texture in severely cold-rolled sheet; aluminium had the original deformation texture, while in Ni–Fe, after recrystallization at 1000° C., the (100) [001] texture was predominant. Both rolling and recrystallization textures varied with depth below the surface, the cubic texture being more perfect near the surface, and also more perfect in a thicker specimen. In Ni–Fe there were, besides grains in the cubic orientation, others in orientations varying with temperature of annealing.

It has been found that previous workers have described the texture of rolled molybdenum too simply.[206] After straight rolling, the texture was now described as a mixture of (1) the ideal (100) [110] texture with a certain spread about the rolling direction at the limit of which [111] became perpendicular to the rolling plane, and (2) a complete rotation about this limiting [111] direction. After cross-rolling, a second [110] direction was aligned with the second rolling direction so that the spread about the rolling direction disappeared, but the rotary symmetrical [111] orientation was unaffected, and therefore occurred in the cross-rolled texture. After annealing the cross-rolled strip at 1180° C., it contained only the (100) [110] texture. The texture of straight-rolled molybdenum was in good agreement with that of straight-rolled iron, another body-centred cubic metal; the texture of cross-rolled iron is not known.

X-rays have also been used to study the structure of deposited surface layers of metal. For electrolytically deposited nickel, it had been found previously that the crystallographic direction of the fibre axis could vary widely. It has, however, proved possible to find conditions under which the fibre axis [100] was reproducible.[207] When deposited on polished copper, layers from 5μ upwards showed the [100] texture, while on etched copper and etched 'cubic orientation' Ni–Fe strip, the orientation was random for thin layers and developed the [100] texture only for relatively thick layers.

Investigation of electrolytically deposited chromium showed that the brightest deposits had a [111] texture, and were confined to a definite range of temperature and current density. Variation from these conditions resulted in (*a*) an increasing proportion of randomly orientated crystals; and (*b*) a reduction in perfection of alignment of crystallites in the preferred orientation.[208]

In electrolytically deposited zinc the orientations were found to vary profoundly with the solutions used; on iron a [103], a [001], and random orientations were found, and on tin a [111] texture was found.[209] The growth of tin crystals on steel has been studied.[210] Results showed a continuous change in orientation along the direction of growth which was attributed to a continuously changing concentration of iron in solid solution in the tin.

Textures at the surface of cast metals have also been a subject of investigation. In zinc and cadmium the basal plane of the hexagonal unit cell tended to lie parallel to the surface; die castings of aluminium and α/β-brass had random surface textures.[211]

The study of the structure and texture of brasses, formed by interdiffusion of copper and zinc deposited on glass by evaporation, has yielded interesting results. The α, β, γ, δ, and ε phases were all detected at first, but finally only β was present; zinc had [001] and [101] textures, copper had [111] and [100] textures, and the final β-brass had a [110] texture.[212]

Extrusion textures of aluminium and brass rods showed a [001] fibre texture for aluminium, a [110]

texture for β-brass, and a faint but definite [001] texture for α-brass.[213]

NEW TECHNIQUES AND APPARATUS

New designs of camera are constantly being evolved for specific researches. Powder cameras have been made at Oxford, Cambridge,[214] Bangor,[215] and in Germany, which enable photographs to be taken of specimens at high temperatures. In the metallurgical field this is of use in determining the structures of phases which cannot be quenched, for following the courses of phase transformations, and for work on equilibrium diagrams. Some low-temperature cameras have also been in use.[216] Micro-cameras have been developed in America,[217, 218] in Holland,[219] and in England at the Cavendish Laboratory, Cambridge ; by this means very small areas of a sample can be examined, a technique which should have considerable application in the study of metals.

The development of rotating-anode X-ray tubes is continuing ; some tubes have been in operation in this country for a number of years.[220] The resultant reduction in exposure time becomes important in some fields of work on metals, such as the study of very complex structures, the following of a fairly rapid reaction in the solid state, and the study of extra reflections.

Bent-crystal monochromators have been developed, particularly in France.[221, 222] Monochromatic radiation is required in several metallurgical investigations, such as the study of diffuse reflections as in the case of age-hardening, and bent-crystal reflectors result in a focused beam of much greater intensity than that obtained by reflection from a plane ; it is also possible to separate out $K\alpha_1$ and $K\alpha_2$ radiations when investigating shapes of powder lines.

The Geiger counter method of recording X-ray diffraction spots or lines is at present being developed in America,[223-225] and also in England. It should lead to some improvements as against the photographic method, such as increase in sensitivity, reduction in time, increase in range of linear response, and the possibility of obtaining a continuous record of a structural change, which will be of importance in some fields of metallurgy such as the study of diffusion in alloys. The Geiger counter is also of great assistance in the setting up of micro-cameras.

Increased accuracy in lattice parameter determination has come mainly from investigations of possible sources of error, and from new proposals for the extrapolation functions to be used.[68, 69] A new method of mounting the film has been suggested by Straumanis,[65] and an improved technique for use with single crystals has been developed.[66,226]

A considerable field of study which has been opened up during the war is that of the diffuse or extra reflections arising from either dynamic (thermal) or static imperfections in the crystal lattice. Mrs. Lonsdale has reviewed work on the former type of imperfections.[227] It now appears that the only theory that can satisfactorily explain results, especially for metal crystals, is that the diffuse reflections are caused by the elastic thermal vibrations of the atoms and molecules. At a later stage it may become possible to determine coefficients of elasticity by this means.[228] Guinier has reviewed work done on the static imperfections[229] due to mechanical stress, atomic rearrangement (age-hardening) or defects in periodicity.

The study of low-angle scattering has also developed, particularly in France and America. This has its particular application in the determination of particle size in coals, etc., but was also used as an indication of platelet size in the study of age-hardening.[159, 165]

The study of the internal structure of Laue diffraction spots has led to a new ' X-ray microscopy ' for the examination of the distribution of strains in the surface of a crystal, which has been developed particularly by Barrett in America.[230]

A bubble model has been devised by Bragg to illustrate some of the properties of metals, such as processes by which glide takes place by movement of a dislocation, the marked effect of impurities, and the process of cold-work and recrystallization.[231]

The method of neutron diffraction, which has been made possible by the building of atomic piles, and is now in process of development, particularly in America, has already been shown to have applications in research on the structure of alloys. These applications depend upon the fact that, while X-rays are scattered by extra-nuclear electrons and the atomic scattering factor is approximately proportional to the atomic number, neutrons are scattered by the atomic nucleus and the scattering varies in an irregular manner with atomic number.

The first application occurs in the cases where the neutron scattering by a light and a heavy atom is comparable, which gives rise to the possibility of locating the elements hydrogen, boron, etc., in the interstitial alloys mentioned on page 300. In this way NaH and NaD have been shown to have an NaCl structure ; sodium and hydrogen scatter neutrons with opposite phase whereas sodium and deuterium scatter neutrons with the same phase.[232] The second application occurs for metals whose X-ray scattering powers are of practically the same magnitude, but whose neutron scattering powers are very different ; by this means the alloy FeCo has been shown to have an ordered CsCl structure below a certain critical temperature.[233] At present, neutron diffraction methods can be applied only to structures of simple types, and to materials of which fairly large specimens are available, since the techniques so far developed are not comparable in resolving power with the corresponding X-ray techniques.

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References

1. H. Lipson and A. R. Stokes: *Nature*, 1941, vol. 148, p. 437.
2. W. H. Beamer and C. R. Maxwell: *Journal of Chemical Physics*, 1946, vol. 14, p. 569.
3. H. Bose and B. B. Ray: *Indian Journal of Physics*, 1941, vol. 15, p. 389.
4. K. D. Gupta, S. R. Das, and B. B. Ray: *Indian Journal of Physics*, 1941, vol. 15, p. 389.
5. H. Hendus: *Zeitschrift für Physik*, 1942, vol. 119, p. 265.
6. O. S. Edwards and H. Lipson: *Proceedings of the Royal Society*, 1942, series A, vol. 180, p. 268.
7. A. J. C. Wilson: *Proceedings of the Royal Society*, 1942, series A, vol. 180, p. 277.
8. (*a*) J. J. Arlman and R. Kronig: *Physica*, 1943, vol. 10, p. 795.
   (*b*) J. Bouman, J. J. Arlman, and L. L. van Reijen: *Physica*, 1946, vol. 12, p. 353.
9. G. D. Preston: *Proceedings of the Royal Society*, 1939, series A, vol. 172, p. 116.
10. G. Chaudron, J. Benard, and A. Michel: *Comptes Rendus*, 1944, vol. 218, p. 913.
11. E. A. Owen and E. St. J. Williams: *Proceedings of the Physical Society*, 1944, vol. 56, p. 52.
12. J. Benard and J. Talbot: *Comptes Rendus*, 1946, vol. 222, p. 493.
13. E. W. Potter and R. W. Huber: *Physical Review*, 1945, vol. 68, p. 24.
14. C. Squire and A. Kaufmann: *Journal of Chemical Physics*, 1941, vol. 9, p. 673.
15. N. J. Petch: *Journal of The Iron and Steel Institute*, 1942, No. I, p. 111P.
16. H. Lipson and A. M. B. Parker: *Journal of The Iron and Steel Institute*, 1944, No. I, p. 123P.
17. H. Lipson and N. J. Petch: *Journal of The Iron and Steel Institute*, 1940, No. II, p. 95P.
18. Proceedings of Conference on X-Ray Analysis: *Journal of Scientific Instruments*, 1947, vol. 24, p. 16.
19. E. Öhman: *Jernkontorets Annaler*, 1944, vol. 128, p. 13.
20. E. N. Kislyakova: *Zhurnal Fizicheskoy Khimii*, 1943, vol. 17, p. 108.
21. Ya. S. Umanskiy and S. S. Khidekel: *Zhurnal Fizicheskoy Khimii*, 1941, vol. 15, pp. 983 and 997.
22. R. Juza and W. Sachze: *Zeitschrift für anorganische und allgemeine Chemie*, 1943, vol. 251, p. 201.
23. V. Epelbaum and A. Brager: *Acta Physicochimica, U.R.S.S.*, 1940, vol. 13, p. 595.
24. W. Hume-Rothery, G. V. Raynor, and A. T. Little: *Journal of The Iron and Steel Institute*, 1942, No. I, p. 129P.
25. G. E. R. Schulze: *Zeitschrift für Elektrochemie und angewandte physikalische Chemie*, 1939, vol. 45, p. 849.
26. U. Dehlinger and G. E. R. Schulze: *Zeitschrift für Kristallographie, Kristallgeometrie, Kristallphysik, Kristallchemie*, 1940, vol. 102, p. 377.
27. H. J. Wallbaum: *Zeitschrift für Kristallographie, Kristallgeometrie, Kristallphysik, Kristallchemie*, 1940, vol. 103, p. 391; *Archiv für das Eisenhüttenwesen*, 1941, vol. 14, p. 521.
28. R. Vogel: *Ricerca Scientifica ed il Progresso Tecnico nell' Economia Nazionale*, 1943, vol. 14, p. 147.
29. W. Füllig, K. Moeller, and R. Vogel: *Zeitschrift für Metallkunde*, 1942, vol. 34, p. 253.
30. F. Laves: *Naturwissenschaften*, 1943, vol. 31, p. 145.
31. E. Hellner and F. Laves: *Zeitschrift für Kristallographie, Kristallgeometrie, Kristallphysik, Kristallchemie*, 1943, vol. 105, p. 134.
32. H. J. Wallbaum: *Naturwissenschaften*, 1943, vol. 31, p. 91.
33. A. R. Weill: *Nature*, 1943, vol. 152, p. 413.
34. G. Ekwall and A. Westgren: *Arkiv for Kemi, Mineralogi och Geologi*, 1940, vol. 14B, No. 7.
35. H. Perlitz and A. Westgren: *Ibid.*, 1943, vol. 16B, No. 13.
36. H. Perlitz and A. Westgren: *Ibid.*, 1942, vol. 15B, No. 16.
37. O. E. Ullner: *Ibid.*, 1940, vol. 14A, No. 3.
38. O. Nial: *Ibid.*, 1943, vol. 17B, No. 11.
39. W. Haucke: *Zeitschrift für anorganische und allgemeine Chemie*, 1940, vol. 244, p. 17.
40. H. Nowotny: *Zeitschrift für Metallkunde*, 1942, vol. 34, p. 247.
41. F. Endtner and W. Klemm: *Zeitschrift für anorganische und allgemeine Chemie*, 1943, vol. 252, p. 64.
42. C. Guillard and J. Wyart: *Comptes Rendus*, 1944, vol. 219, p. 393.
43. S. Kiuti: *Japan Nickel Review*, 1941, vol. 9, p. 78.
44. E. S. Makarov: *Izvestiya Akademii Nauk, S.S.S.R. (Kemi)*, 1943, part 4, p. 264; 1944, part 29, pp. 114 and 201.
45. F. Laves and H. J. Wallbaum: *Chemisches Zentrallblatt*, 1943, vol. 114, p. 250.
46. H. Perlitz: *Nature*, 1944, vol. 154, p. 606.
47. A. M. B. Parker: *Nature*, 1945, vol. 156, p. 783.
48. H. Wiehr: *Aluminium-Archiv*, 1940, No. 31, 14 pp.
49. W. Klemm and N. Fratini: *Zeitschrift für anorganische und allgemeine Chemie*, 1943, vol. 251, p. 222.
50. V. Montoro: *Metallurgia Italiana*, 1943, vol. 35, p. 57.
51. H. J. Wallbaum: *Zeitschrift für Metallkunde*, 1943, vol. 35, p. 200.
52. G. Brauer: *Zeitschrift für Elektrochemie und angewandte physikalische Chemie*, 1943, vol. 49, p. 208.
53. D. Harker: *Journal of Chemical Physics*, 1944, vol. 12, p. 315.
54. H. Nowotny and K. Schubert: *Naturwissenschaften*, 1943, vol. 31, p. 592.
55. H. Nowotny: *Zeitschrift für Metallkunde*, 1942, vol. 34, p. 237.
56. A. Iandelli: *Gazzetta Chimica Italiana*, 1941, vol. 71, p. 58.
57. H. Nowotny: *Zeitschrift für Metallkunde*, 1942, vol. 34, p. 22.
58. W. Hofmann and H. Wiehr: *Ibid.*, 1941, vol. 33, p. 369.
59. G. Brauer and R. Rudolph: *Zeitschrift für anorganische und allgemeine Chemie*, 1941, vol. 248, p. 405.
60. W. Hofmann: *Zeitschrift für Metallkunde*, 1941, vol. 33, p. 373.
61. H. Nowotny: *Ibid.*, 1941, vol. 33, p. 388.
62. A. Osawa and N. Shibata: *Nippon Kinzoku Gakkai-Si*, 1940, vol. 4, p. 362.
63. F. Laves and H. J. Wallbaum: *Naturwissenschaften*, 1939, vol. 27, p. 674.
64. H. Lipson and A. Taylor: *Proceedings of the Royal Society*, 1939, series A, vol. 173, p. 232.
65. M. Straumanis: *Zeitschrift für Kristallographie, Kristallgeometrie, Kristallphysik, Kristallchemie*, 1940, vol. 102, p. 432.
66. M. C. M. Farquhar and H. Lipson: *Proceedings of the Physical Society*, 1946, vol. 58, p. 200.
67. H. Lipson and A. J. C. Wilson: *Journal of Scientific Instruments*, 1941, vol. 18, p. 144.
68. J. B. Nelson and D. P. Riley: *Proceedings of the Physical Society*, 1945, vol. 57, p. 160.
69. A. Taylor and H. Sinclair: *Ibid.*, 1945, vol. 57, p. 126.
70. A. R. Stokes and A. J. C. Wilson: *Ibid.*, 1941, vol. 53, p. 658.
71. A. J. C. Wilson: *Ibid.*, 1942, vol. 54, p. 487.
72. A. Gott: *Annales de Physique*, 1942, vol. 41, p. 520.
73. W. Hume-Rothery and K. W. Andrews: *Journal of the Institute of Metals*, 1942, vol. 68, p. 19.
74. K. Lonsdale and W. Hume-Rothery: *The London, Edinburgh, and Dublin Philosophical Magazine and Journal of Science*, 1945, vol. 36, p. 799.

75. K. LONSDALE and W. HUME-ROTHERY : *Ibid.*, 1945, vol. 36, p. 842.
76. E. A. OWEN and Y. H. LIU : *Ibid.*, 1947, vol. 38, p. 354.
77. G. SIEBEL and H. VOSSKÜHLER : *Zeitschrift für Metallkunde*, 1939, vol. 31, p. 359.
78. V. I. MIKHEEVA : *Izvestiya Akademii Nauk S.S.S.R.*, 1940, vol. 5, p. 775.
79. N. W. AGEEW and N. YA. TALYZIN : *Izvestiya Sektora Fiziko-Khimicheskogo Analiza*, 1940, vol. 13, p. 251.
80. W. HUME-ROTHERY and G. V. RAYNOR : *Proceedings of the Royal Society*, 1940, series A, vol. 177, p. 27.
81. G. V. RAYNOR : *Ibid.*, 1940, series A, vol. 174, p. 457.
82. G. V. RAYNOR : *Ibid.*, 1942, series A, vol. 180, p. 107.
83. F. FOOTE and E. R. JETTE : *Transactions of the American Institute of Mining and Metallurgical Engineers*, 1941, vol. 143, p. 124.
84. A. J. KING : *Journal of the American Chemical Society*, 1942, vol. 64, p. 1226.
85. H. LIPSON, N. J. PETCH, and D. STOCKDALE : *Journal of the Institute of Metals*, 1941, vol. 67, p. 79.
86. F. E. BOWMAN and others : *Transactions of the American Society for Metals*, 1943, vol. 31, p. 487 ; 1944, vol. 33, p. 481 ; 1945, vol. 35, p. 112.
87. M. C. M. FARQUHAR, H. LIPSON, and A. R. WEILL : *Journal of The Iron and Steel Institute*, 1945, vol. 152, p. 457.
88. K. W. ANDREWS and W. HUME-ROTHERY : *Proceedings of the Royal Society*, 1941, series A, vol. 178, p. 464.
89. E. A. OWEN : *Journal of the Institute of Metals*, 1946, vol. 73, p. 471.
90. C. S. BARRETT : *Journal of Applied Physics*, 1941, vol. 12, p. 385.
91. W. HUME-ROTHERY and G. V. RAYNOR : *Journal of Scientific Instruments*, 1941, vol. 18, p. 74.
92. H. LIPSON : *Journal of the Institute of Metals*, 1943, vol. 69, p. 1.
93. A. OSAWA and T. MURATA : *Nippon Kinzoku Gakkai-Si*, 1940, vol. 4, p. 228.
94. W. C. ELLIS and E. S. GREINER : *Transactions of the American Society for Metals*, 1941, vol. 29, p. 415.
95. W. F. EHRET and D. H. GURINSKY : *Journal of the American Chemical Society*, 1943, vol. 65, p. 1226.
96. R. HULTGREN and C. A. ZAPFFE : *Transactions of the American Institute of Mining and Metallurgical Engineers*, 1939, vol. 133, p. 58.
97. A. R. TROIANO and F. T. MCGUIRE : *Transactions of the American Society for Metals*, 1943, vol. 31, p. 340.
98. B. L. AVERBACH : *Metals and Alloys*, 1941, vol. 13, p. 730 ; 1941, vol. 14, p. 47.
99. R. S. DEAN and others : *Transactions of the American Society for Metals*, 1945, vol. 34, p. 442.
100. A. J. BRADLEY and H. J. GOLDSCHMIDT : *Journal of The Iron and Steel Institute*, 1939, No. II, p. 11P.
101. E. A. OWEN and A. H. SULLY : *The London, Edinburgh, and Dublin Philosophical Magazine*, 1941, vol. 31, p. 314.
102. H. J. WALLBAUM : *Archiv für das Eisenhüttenwesen*, 1941, vol. 14, p. 521.
103. F. GÖTZL, F. HALLA, and J. SCHRAMM : *Zeitschrift für Metallkunde*, 1941, vol. 33, p. 375.
104. J. SCHRAMM : *Ibid.*, 1941, vol. 33, p. 381.
105. E. GEBHARDT and W. KÖSTER : *Ibid.*, 1940, vol. 32, p. 253.
106. R. HULTGREN and R. I. JAFFÉ : *Journal of Applied Physics*, 1941, vol. 12, p. 501.
107. T. HEUMANN : *Zeitschrift für Metallkunde*, 1943, vol. 35, p. 206.
108. F. H. ELLINGER : *Transactions of the American Society for Metals*, 1942, vol. 30, p. 607.
109. J. SCHRAMM : *Zeitschrift für Metallkunde*, 1941, vol. 33, p. 347.
110. F. H. ELLINGER and W. P. SYKES : *Transactions of the American Society for Metals*, 1940, vol. 28, p. 619.
111. A. G. H. ANDERSON : *Transactions of the American Institute of Mining and Metallurgical Engineers*, 1940, vol. 137, p. 334.
112. T. ISAWA : *Nippon Kinzoku Gakkai-Si*, 1940, vol. 4, p. 398.
113. F. W. JONES and C. SYKES : *Journal of the Institute of Metals*, 1939, vol. 65, p. 419.
114. N. W. AGEEW and E. S. MAKAROV : *Izvestiya Sektora Fiziko-Khimicheskogo Analiza*, 1940, vol. 13, p. 171.
115. E. A. OWEN, J. ROGERS, and J. C. GUTHRIE : *Journal of the Institute of Metals*, 1939, vol. 65, p. 457.
116. K. W. ANDREWS and others : *Proceedings of the Royal Society*, 1941, series A, vol. 177, p. 149.
117. H. R. LETNER, H. H. GEORGE, and S. S. SIDHU : *Physical Review*, 1940, vol. 58, p. 206.
118. A. SCHNEIDER and U. ESCH : *Zeitschrift für Elektrochemie und angewandte physikalische Chemie*, 1943, vol. 49, p. 72.
119. E. A. OWEN and E. A. O'D. ROBERTS : *Journal of the Institute of Metals*, 1945, vol. 71, p. 213.
120. R. I. JAFFÉ, E. M. SMITH, and B. W. GONSER : *Transactions of the American Institute of Mining and Metallurgical Engineers*, 1945, vol. 161, p. 366.
121. E. A. OWEN and E. A. O'D. ROBERTS : *Journal of the Institute of Metals*, 1940, vol. 66, p. 389.
122. E. A. OWEN and W. H. REES : *Journal of the Institute of Metals*, 1941, vol. 67, p. 141.
123. O. NIAL : Inaugural Dissertation, University of Stockholm, 1945.
124. J. SCHRAMM : *Zeitschrift für Metallkunde*, 1940, vol. 32, p. 399.
125. W. KLEMM and G. MIKA : *Zeitschrift für anorganische und allgemeine Chemie*, 1941, vol. 248, p. 155.
126. H. NOWOTNY and H. WESENBERG : *Zeitschrift für Metallkunde*, 1939, vol. 31, p. 363.
127. W. HUME-ROTHERY and G. V. RAYNOR : *Proceedings of the Royal Society*, 1940, series A, vol. 174, p. 471.
128. N. S. KURNAKOV and V. I. MIKLEEVA : *Izvestiya Sektora Fiziko-Khimicheskogo Analiza*, 1940, vol. 13, p. 209.
129. E. GEBHARDT and W. KOSTER : *Zeitschrift für Metallkunde*, 1940, vol. 32, p. 262.
130. S. VALENTINER : *Ibid.*, 1940, vol. 32, p. 244.
131. F. H. ELLINGER : *Transactions of the American Society for Metals*, 1943, vol. 31, p. 89.
132. A. J. BRADLEY, W. L. BRAGG, and C. SYKES : *Journal of The Iron and Steel Institute*, 1940, No. I, p. 63P.
133. G. V. RAYNOR and K. LITTLE : *Journal of the Institute of Metals*, 1945, vol. 71, p. 493.
134. F. ERDMANN-JESNITZER : *Aluminium-Archiv*, 1940, No. 29, 14pp.
135. V. G. KUZNETSOV and E. S. MAKAROV : *Izvestiya Sektora Fiziko-Khimicheskogo Analiza*, 1940, vol. 13, p. 177.
136. E. DEGISCHER: *Aluminium-Archiv*, 1939, No. 18, 19 pp.
137. O. S. EDWARDS: *Journal of the Institute of Metals*, 1941, vol. 67, p. 67.
138. V. G. KUZNETSOV and L. N. GUSEVA : *Izvestiya Akademii Nauk S.S.S.R.*, 1940, vol. 6, p. 905.
139. W. GUERTLER and G. RASSMANN : *Metallwirtschaft*, 1943, vol. 122, p. 1.
140. R. VOGEL and T. HEUMANN : *Archiv für das Eisenhüttenwesen*, 1944, vol. 17, p. 271.
141. R. VOGEL and F. HILLEN : *Ibid.*, 1942, vol. 15, p. 551.
142. R. VOGEL and B. WENDEROTT : *Ibid.*, 1941, vol. 14, p. 279.
143. R. VOGEL and C. JENTZCH-USCHINSKI : *Ibid.*, 1940, vol. 13, p. 403.
144. A. J. BRADLEY and H. J. GOLDSCHMIDT : *Journal of The Iron and Steel Institute*, 1941, No. II, p. 273P.
145. A. J. BRADLEY, W. F. COX, and H. J. GOLDSCHMIDT : *Journal of the Institute of Metals*, 1941, vol. 67, p. 189.

146. E. S. GREINER and E. R. JETTE : *Transactions of the American Institute of Mining and Metallurgical Engineers*, 1943, vol. 152, p. 48.
147. V. G. KUZNETSOV and N. N. EVSEEVA : *Zhurnal Prikladnoy Khimii*, 1939, vol. 12, p. 406.
148. J. SCHRAMM : *Zeitschrift für Metallkunde*, 1941, vol. 33, p. 403.
149. T. MURAKAMI and others : *Nippon Kirzoku Gakkai-Si*, 1940, vol. 4, p. 189.
150. A. G. H. ANDERSON and A. W. KINGSBURY : *Transactions of the American Institute of Mining and Metallurgical Engineers*, 1943, vol. 152, p. 38.
151. N. SHIBATA : *Nippon Kirzoku Gakkai-Si*, 1940, vol. 4, p. 269 ; 1941, vol. 5, p. 12.
152. A. YUKATA : *Ibid.*, 1941, vol. 5, p. 136.
153. (*a*) K. IWASE and M. OKAMOTO : *Ibid.*, 1941, vol. 5, p. 82.
(*b*) M. OKAMOTO : *Ibid.*, 1939, vol. 3, p. 44.
154. M. OKAMOTO : *Ibid.*, 1939, vol. 3, pp. 336 and 365.
155. K. MOELLER : *Zeitschrift für Metallkunde*, 1942, vol. 34, p. 371.
156. K. MOELLER : *Ibid.*, 1943, vol. 35, p. 27.
157. P. LACOMBE and G. CHAUDRON : *Comptes Rendus*, 1939, vol. 209, p. 306.
158. G. D. PRESTON : *Journal of Scientific Instruments*, 1941, vol. 18, p. 154.
159. A. GUINIER : *Journal de Physique et le Radium*, 1942, vol. 3, p. 124.
160. A. GUINIER : *Mésures*, 1946, Oct.
161. C. S. TOWNES : *Transactions of the American Society for Metals*, 1944, vol. 33, p. 281.
162. C. S. BARRETT, A. H. GEISLER, and R. F. MEHL : *Transactions of the American Institute of Mining and Metallurgical Engineers*, 1941, vol. 143, p. 134.
163. A. H. GEISLER, C. S. BARRETT, and R. F. MEHL : *Ibid.*, 1943, vol. 152, p. 201.
164. P. LACOMBE : *Comptes Rendus*, 1943, vol. 217, p. 175.
165. A. GUINIER : *Métaux et Corrosion*, 1943, vol. 18, p. 209.
166. A. GUINIER and P. JACQUET : *Comptes Rendus*, 1943, vol. 217, p. 22.
167. D. HARKER : *Transactions of the American Society for Metals*, 1944, vol. 32, p. 210.
168. C. S. SMITH : *Journal of Applied Physics*, 1941, vol. 12, p. 817.
169. E. SCHEUFELE : *Zeitschrift für Metallkunde*, 1941, vol. 33, p. 214.
170. F. W. JONES, P. LEECH, and C. SYKES : *Proceedings of the Royal Society*, 1943, series A, vol. 181, p. 154.
171. E. NABARRO : *Proceedings of the Physical Society*, 1940, vol. 52, p. 90 ; *Proceedings of the Royal Society*, 1940, series A, vol. 175, p. 519.
172. V. DANIEL and H. LIPSON : *Proceedings of the Royal Society*, 1943, series A, vol. 181, p. 368 ; 1944, series A, vol. 182, p. 378.
173. A. B. GRENINGER and A. R. TROIANO : *Transactions of the American Institute of Mining and Metallurgical Engineers*, 1941, vol. 145, p. 289 ; 1940, vol. 140, p. 307.
174. G. V. SMITH and R. F. MEHL : *Ibid.*, 1942, vol. 150, p. 211.
175. M. ARBUSOW and G. KURDJUMOW : *Journal of Physics* (*U.S.S.R.*), 1942, vol. 6, p. 163.
176. A. KOMAR and T. TARASOV : *Zhurnal Tekhnicheskoy Fiziki*, 1941, vol. 11, p. 1087.
177. S. WOO, C. S. BARRETT, and R. F. MEHL : *Transactions of the American Institute of Mining and Metallurgical Engineers*, 1944, vol. 156, p. 100.
178. R. S. DEAN and others : *Transactions of the American Society for Metals*, 1945, vol. 34, p. 465.
179. A. HARGREAVES : *Journal of Scientific Instruments*, 1944, vol. 21, p. 118.
180. A. GUINIER : Report on X-Ray Department, 1940–44, Laboratoire D'Essais, Conservatoire National des Arts et Métiers, Paris.
181. W. A. WOOSTER : *Journal of Scientific Instruments*, 1948, vol. 25, p. 129.
182. C. S. BARRETT and L. H. LEVENSON : *Transactions of the American Institute of Mining and Metallurgical Engineers*, 1941, vol. 145, p. 281.
183. C. S. BARRETT and F. W. STEADMAN : *Ibid.*, 1942, vol. 147, p. 57.
184. M. COOK and T. LL. RICHARDS : *Journal of the Institute of Metals*, 1940, vol. 66 p. 1 ; 1941, vol. 67, p. 203.
185. M. COOK and T. LL. RICHARDS : *Ibid.*, 1944, vol. 70, p. 159.
186. H. C. VACHER : *Journal of Research of the National Bureau of Standards*, 1941, vol. 26, p. 385.
187. R. M. BRICK and M. A. WILLIAMSON : *Transactions of the American Institute of Mining and Metallurgical Engineers*, 1941, vol. 143, p. 84.
188. P. COHEUR : *Revue Universelle des Mines, de la Métallurgie, des Travaux Publics, des Sciences et des Arts Appliqués à l'Industrie*, 1943, vol. 6, p. 1.
189. R. M. BRICK : *Transactions of the American Institute of Mining and Metallurgical Engineers*, 1940, vol. 137, p. 193.
190. A. E. BRYUKHANOV : *Zhurnal Tekhnicheskoy Fiziki*, 1939, vol. 9, p. 1745.
191. M. COOK and T. LL. RICHARDS : *Journal of the Institute of Metals*, 1943, vol. 69, p. 351.
192. P. COHEUR and C. S. BARRETT : *Transactions of the American Institute of Mining and Metallurgical Engineers*, 1947, vol. 171, p. 63.
193. F. H. WILSON and R. M. BRICK : *Ibid.*, 1945, vol. 161, p. 173.
194. E. E. SPILLETT : *Journal of the Institute of Metals*, 1943, vol. 69, p. 149.
195. W. KISTLER : *Schweizer Archiv für angewandte Wissenschaft und Technik*, 1946, vol. 12, p. 169.
196. D. E. THOMAS : *The London, Edinburgh, and Dublin Philosophical Magazine and Journal of Science*, 1941, vol. 31, p. 425.
197. P. W. BAKARIAN : *Transactions of the American Institute of Mining and Metallurgical Engineers*, 1942, vol. 147, p. 266.
198. A. HARGREAVES : *Journal of the Institute of Metals*, 1945, vol. 71, p. 73.
199. C. S. BARRETT and C. T. HALLER : *Transactions of the American Institute of Mining and Metallurgical Engineers*, 1947, vol. 171, p. 246.
200. J. T. BURWELL : *Metals Technology*, 1940, vol. 7, Feb. Technical Publication No. 1178.
201. N. P. GOSS : *Transactions of the American Society for Metals*, 1941, vol. 29, p. 20.
202. C. G. DUNN : *Transactions of the American Institute of Mining and Metallurgical Engineers*, 1945, vol. 161, p. 98.
203. G. SACHS and J. SPRETNAK : *Metals Technology*, 1940, vol.7, Jan. Technical Publication, No. 1143.
204. J. F. H. CUSTERS and G. W. RATHENAU : *Physica*, 1941, vol. 8, pp. 759 and 771.
205. J. F. H. CUSTERS and G. W. RATHENAU : *Physica*, 1947, vol. 13, pp. 97 and 336.
206. J. F. H. CUSTERS and J. C. RIERSMA : *Physica*, 1946, vol. 12, p. 195.
207. W. G. BURGERS and H. E. QUANJEL : *Recueil des Travaux Chimiques des Pays-Bas*, 1942, vol. 61, p. 353.
208. W. HUME-ROTHERY and M. R. J. WYLIE : *Proceedings of the Royal Society*, 1943, series A, vol. 181, p. 331.
209. P. I. USIKOV : *Zhurnal Prikladnoy Kimii*, 1939, vol. 12, p. 476.
210. B. CHALMERS : *Nature*, 1940, vol. 146, p. 493.
211. G. EDMUNDS : *Transactions of the American Institute of Mining and Metallurgical Engineers*, 1945, vol. 161, p. 114.
212. A. A. BURR, H. S. COLEMAN, and W. P. DAVEY : *Transactions of the American Society for Metals*, 1944, vol. 33, p. 73.
213. H. UNCKEL : *Zeitschrift für Metallkunde*, 1940, vol. 32, p. 343.
214. A. J. C. WILSON : *Proceedings of the Physical Society*, 1941, vol. 53, p. 235.

215. E. A. OWEN : *Journal of Scientific Instruments*, 1943, vol. 20, p. 190.
216. W. HUME-ROTHERY and D. J. STRAWBRIDGE : *Ibid.*, 1947, vol. 24, p. 89.
217. I. FANKUCKEN and H. MARK : *Journal of Applied Physics*, 1944, vol. 15, p. 364.
218. F. E. CHESLEY : *Review of Scientific Instruments*, 1947, vol. 18, p. 422.
219. D. KREGER : *Proceedings of Koninklijke Nederlandsche Akademie van Wetenschappen*, 1946, vol. 48, p. 336.
220. I. MACARTHUR : *Electronic Engineering*, 1944, vol. 16, p. 272 ; 1945, vol. 17, p. 317.
221. A. GUINIER : *Annales de Physique*, 1939, vol. 12, p. 161.
222. J. BARRAUD : *Comptes Rendus*, 1942, vol. 214, p. 795.
223. H. FRIEDMAN : *Electronics*, 1945, vol. 18, Apr., p. 132.
224. J. BLECKSMA, G. KLOOS, and H. J. DI GIOVANNI : *Philips Technical Review*, 1948, vol. 10, p. 1.
225. B. F. DECKER, E. T. ASP, and D. HARKER : *Journal of Applied Physics*, 1948, vol. 19, p. 388.
226. O. WEISZ, W. COCHRAN, and W. F. COLE : *Acta Crystallographica*, 1948, vol. 1, p. 83.
227. K. LONSDALE : *Proceedings of the Physical Society*, 1942, vol. 54, p. 314.
228. J. LAVAL : *Comptes Rendus*, 1942, vol. 214, p. 623.
229. A. GUINIER : *Proceedings of the Physical Society*, 1945, vol. 57, p. 310.
230. C. S. BARRETT : *Transactions of the American Institute of Mining and Metallurgical Engineers*, 1945, vol. 161, p. 15.
231. W. L. BRAGG : *Journal of Scientific Instruments*, 1942, vol. 19, p. 148.
232. C. G. SHULL and others : *Physical Review*, 1948, vol. 73, p. 842.
233. C. G. SHULL and E. O. WOLLAN : *Science*, 1948, vol. 108, p. 69.

Part II—THERMAL CONDUCTIVITIES OF MOLTEN METALS AND ALLOYS

By R. W. Powell, D.Sc., Ph.D., F.Inst.P.

SYNOPSIS

Available data for the thermal conductivities of molten metals and alloys are presented graphically and in tabular form. These values are discussed from several theoretical aspects, including that involving their relations with the electrical conductivites of the molten metals. This examination leads to estimates being made giving approximate values for the thermal conductivities of other molten metals not yet tested experimentally.

It appears that a very definite need exists on both practical and theoretical grounds for further experiments to be conducted in this field.

Data on the atomic heats of molten metals are included as an Appendix.

FROM time to time information is required regarding the properties which determine the relative efficiencies of molten metals for use as convective coolants. Steel quenching may be regarded as one aspect of this problem. It is true that molten metals may not be employed to any considerable extent in normal steel processing, but there have been some special applications as, for instance, the use of a bath of molten lead in the course of **S**-curve determinations. The main properties concerned are viscosity, density, atomic heat, and thermal conductivity. The present paper deals mainly with the last-mentioned property, and owing to the relatively small amount of information available for the thermal conductivities of molten metals, the survey has not been restricted to those molten metals which are likely to be of practical value.

Examination of the literature has revealed that the thermal conductivities of only ten metals and five alloys have been determined for both solid and liquid states. Of these, only six of the metals have been studied in the liquid state by more than one observer. Figure 1 contains the results obtained to date for the thermal conductivities of liquid metals and alloys. It seems evident that there is need for further work, not only to provide data for other liquid metals and alloys, but also to confirm and extend the results for those already studied. Reference is made to the related property of electrical conductivity and to certain theoretical aspects. Probable values are estimated for the thermal conductivities of some liquid metals which have not yet been examined experimentally. These sections are included to show that the subject also embraces matters of academic interest which are likely to be clarified as a result of further experimental work.

An Appendix contains values for the atomic heats of molten metals.

METALS FOR WHICH INFORMATION IS AVAILABLE

Thermal conductivity values, expressed as calories per square centimetre per second for 1-cm. thickness and 1° C. difference in temperature, which have been obtained for each molten metal, are given in the following pages. Where the same observer has also studied the thermal conductivity in the solid state, these values are included and are discussed in relation to other results for the solid phase of the metal. In general the thermal conductivity undergoes an abrupt change at the melting point. The data include values of K_S/K_L, the ratio of the thermal conductivity of the solid to that of the liquid metal, the values of K_S and K_L being obtained by extrapolation to the

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Dr. Powell is at the Physics Division of the National Physical Laboratory, Teddington, Middlesex.

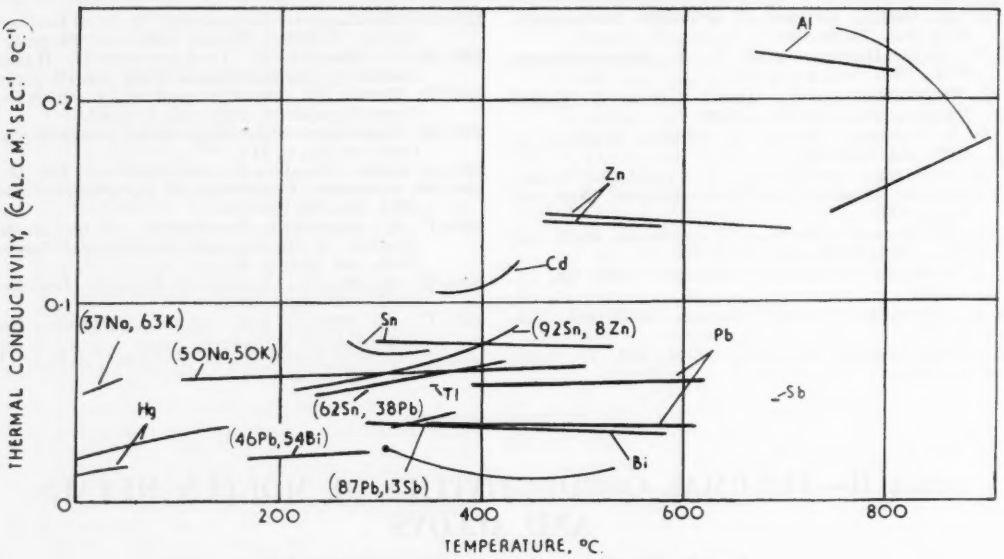

Fig. 1—Thermal conductivity of molten metals

melting point. Often, however, there is a fair amount of possible latitude in making these extrapolations. Values of the corresponding ratio for the electrical conductivity of the solid, σ_S, to that of the liquid, σ_L, are given when they have been obtained in the course of the same investigation, and are referred to in the discussion when they have been determined independently.

It is shown that there is a need for further investigation of both thermal and electrical conductivities of metals, not only for the molten state, but also on transition from the solid to the liquid phase.

Thermal Conductivity of Aluminium (Melting point* 660·0° C.)

Only Konno[1] and Bidwell and Hogan[2] have studied solid and liquid aluminium. In Konno's work the determination was made relative to Lees' value for the thermal conductivity of aluminium at 18° C. The results differ in that whereas Lees had obtained a positive temperature coefficient from — 78° C. to 18° C., Konno obtained a negative coefficient from 18° C. upwards. At a later date Bailey[4] made measurements of the thermal conductivity of Lees' specimen to higher temperatures. He found the positive coefficient to persist to a maximum conductivity value of 0·538 at 250° C., and only thereafter was a negative coefficient obtained. Bailey's value at 550° C., the upper limit of his investigation, is 29% higher than Konno's value at 550° C., and 9% higher than Bidwell and Hogan's curve at that temperature. Whilst their curve for the solid state resembles Konno's in that it has a negative temperature coefficient throughout the range, several other workers have obtained positive coefficients at normal temperatures. The two sets of data available for the liquid state differ appreciably, both as regards absolute values and the sign of the temperature coefficient.

Konno[1]		*Bidwell and Hogan*[2]	
Temp., ° C.	Thermal Cond.	Temp., ° C.	Thermal Cond.
Solid		**Solid**	
18	0·504*	25	0·538
273	0·471	250	0·486
430	0·425	450	0·452
605	0·360	650	0·445
Liquid		**Liquid**	
675	0·223	740	0·143
800	0·214	900	0·180
$K_S/K_L = 1·47$		$K_S/K_L = 3·7$	

* Value due to Lees,[3] assumed by Konno.

The values obtained by Northrup,[5] Tsutsumi,[6] and Matuyama,[7] for the ratio of the electrical conductivities, σ_S/σ_L, are 1·6, 1·65, and 2·2 respectively, which are all greater than Konno's value for K_S/K_L, but much less than Bidwell and Hogan's value.

There is certainly need for further work on the thermal conductivity of molten aluminium.

Thermal Conductivity of Antimony (Melting point 630·5° C.)

For the thermal conductivity of antimony, values available at 0° C. range from 0·038 to 0·053, so that the initial value assumed by Konno represents a fair mean. Also his negative coefficient is very close to that of Lorenz over the range 0–100° C. No values above 100° C. are available for comparison.

Konno[1]			
Temp., ° C.	Thermal Cond.	Temp., ° C.	Thermal Cond.
Solid		**Liquid**	
0	0·0442*	692	0·0503
182	0·0386	$K_S/K_L = 1·1$	
344	0·0414		
469	0·0456		
557	0·0510		
610	0·0575		

* Value due to Lorenz,[8] assumed by Konno.

* The melting points quoted are taken from "Tables of Physical and Chemical Constants," by G. W. C. Kaye and T. H. Laby.

It will be seen that Konno's results indicate a minimum value for the thermal conductivity at about 200° C. and show a small decrease on passing from the solid to the liquid phase. On the other hand, it seems to be established that the change of phase is accompanied by a definite increase in the electrical conductivity, since a value for σ_S/σ_L of the order of 0·7 has been reported from four sources.[5–7, 9] This apparent difference in the behaviour of the thermal and electrical conductivities of antimony in the neighbourhood of the melting point is another result requiring further investigation.

Thermal Conductivity of Bismuth (Melting point 271·3 ± 0·1° C.)

The values for the thermal conductivity of bismuth given below appear to be the only results available above 160° C. Konno's results below this temperature are in reasonable agreement with those of other workers. Regarding the values nearer the liquid state, greater weight should be placed on Konno's values as Northrup and Pratt's method appears to be less accurate ; also, the conductivity of their comparison steel is not given and a likely value has been assumed from their statement that the steel was from a cold-rolled steel shafting. The two values for K_S agree fairly well, but those for K_L differ appreciably, a much greater increase in thermal conductivity on melting being found by Konno :

Konno[1]		*Northrup and Pratt*[10]	
Temp., ° C.	Thermal Cond.	Temp., ° C.	Thermal Cond.
Solid		**Solid**	
18	0·0194*	181	0·015†
89	0·0181	217	$0·014_5$†
160	0·0170	237	0·017†
222	0·0177		
256	0·0183		
Liquid		**Liquid**	
286	0·0400	306	$0·026_5$†
376	0·0378		
484	0·0372		
584	0·0369		
$K_S/K_L = 0·45$		$K_S/K_L = 0·6_5$	

* Value due to Jaeger and Diesselhorst,[11] assumed by Konno.

† Northrup and Pratt's results were expressed in terms of the thermal resistivity compared with steel. The values given in the above are obtained by assuming thermal conductivity values of the order of $0·12–0·11_5$ for the steel over the temperature range 140–200° C.

Several observers have studied the change in electrical conductivity, and the values obtained for σ_S/σ_L have ranged from $0·43_5$ to 0·59, so that there seems to be a fairly close similarity between the behaviour of the thermal and electrical conductivities.

Thermal Conductivity of Cadmium (Melting point 320·9° C).

Brown's results for the thermal conductivity of solid cadmium differ from those of the five other investigators who have made measurements on this metal, in that Brown's are the only results giving a positive temperature coefficient. The general agreement at normal temperatures is fairly good, but owing to the difference in sign of the temperature coefficient, considerable variation occurs nearer the melting point. Thus at 300° C., Brown's results extrapolate to a value of 0·25, whereas Bailey obtained a value of 0·204.

Brown[12]

Temp., ° C.	Thermal Cond.
Solid	
45	0·200
99	0·209
125	0·213
165	0·221
221	0·237
240	0·239
Liquid	
355	0·106
358	0·105
380	0·105
435	0·119

$K_S/K_L = 2·4$

The value of 2·4 which Brown obtained for K_S/K_L is unusually large and is much greater than the values obtained by other investigators for σ_S/σ_L, which ranged from 1·78 to 1·97. For cadmium there is again need for further determinations of thermal conductivity to be made for both solid and liquid phases.

Thermal Conductivity of Lead (Melting point 327·4° C.)

Lead is another metal for which the results of two workers are available for the thermal conductivity in the molten state. It will be seen that despite the excellent agreement over the whole range for the solid state, Bidwell's values for the thermal conductivity of molten lead are about 50% greater than Konno's, and have temperature coefficients of opposite sign :

Konno[1]		*Bidwell*[13]*	
Temp., ° C.	Thermal Cond.	Temp., ° C.	Therma Cond.
Solid		**Solid**	
18	0·083†	30	0·083
108	0·080	90	0·080
222	0·077	250	0·075
298	0·074		
326	0·069		
Liquid		**Liquid**	
355	0·039	390	0·059
447	0·038	445	0·059
531	0·037	620	0·060
601	0·037		
$K_S/K_L = 1·83$		$K_S/K_L = 1·24$	

* Bidwell's values have been read from his published curve.

† Value due to Lees,[3] assumed by Konno.

Several determinations have been made of the ratio σ_S/σ_L, the values obtained ranging from 1·65 to 2·06, with a mean of 1·98, so that, if the thermal and electrical conductivities behave in a similar manner, these results would tend to support Konno's value for K_S/K_L ; Konno's values for the molten state are therefore considered the more probable.

There is clearly need for a re-determination of the thermal conductivity of molten lead.

Thermal Conductivity of Mercury (Melting point −38·86° C.)

More investigations have been made of the thermal conductivity of mercury for the liquid than for the solid state. It will be seen from the following data

that the five observers who have limited their attention to the liquid state obtained values which indicate that the thermal conductivity at 20° C. is of the order of 0·017 to 0·021 and has an appreciable temperature coefficient. Gehlhoff and Neumeier, who appear to be the only workers who have made measurements on both solid and liquid phases, differ in that their results for liquid mercury lie on a much higher, although roughly parallel, line. At 20° C. their value is 0·027.

Observer	Temp., ° C.	Thermal Cond.
	Solid	
Gehlhoff and Neumeier[14]	−193	0·116
,, ,, ,,	−115·5	0·0929
,, ,, ,,	−78·4	0·0776
,, ,, ,,	−44·2	0·0664
	Liquid	
,, ,, ,,	−37·2	0·0218
,, ,, ,,	0·0	0·0248
H. F. Weber[16]	0·0	0·0148
Nettleton[20]	15·5	0·0201
R. Weber[19]	17·0	0·0197
H. F. Weber[16]	50·0	0·0189
Angstrom[17]	50·0	0·0177
Berget[18]	50·0	0·02015
Gehlhoff and Neumeier[14]	50·4	0·0298
,, ,, ,,	99·9	0·0349
,, ,, ,,	149·4	0·0385

Gehlhoff and Neumeier[14] $K_S/K_L = 2·95$, $\sigma_S/\sigma_L = 3·8$
Porter and Simeon[15] $K_S/K_L = 3·91$

For the solid state, comparison can be made only with the results of Reddemann,[21] who studied the conductivity of single crystals of mercury at − 187° and − 77° C. The results at these two temperatures for the direction parallel to the crystal axis were 0·095 and 0·081, and those for the perpendicular direction were 0·069 and 0·063. From the above data it will be seen that Gehlhoff and Neumeier's values agree more closely with Reddemann's data for the direction parallel to the crystal axis. This is the direction for which K_S/K_L would appear to have the larger value. Gehloff and Neumeier's value for σ_S/σ_L agrees with Porter and Simeon's value for K_S/K_L, whilst their own value for K_S/K_L is much lower. Closer agreement would be obtained if their value for K_L had been lower and in line with the data of the other observers quoted above. On the other hand, values of σ_S/σ_L have ranged from 2·2 to 5·0 and are probably dependent on the crystal form of the solid metal.

There is clearly need for further determinations of the thermal conductivity of liquid mercury, and as this metal is readily available and is in the liquid state at normal temperatures it is one which could most readily be included in any fresh investigation into the thermal conductivity of molten metals.

Thermal Conductivity of Sodium (Melting point 97·7 ± 0·2° C.)

No actual values have been published for the thermal conductivity of liquid sodium. Porter and Simeon[15] obtained a value of 1·31 for the ratio K_S/K_L, and a value of 1·438 for the ratio σ_S/σ_L. Other observers have found values ranging from 1·34 to 1·7 for the latter ratio, and Porter and Simeon concluded that both conductivities suffered comparable changes on melting. Determinations of the thermal conductivity of solid sodium have been made by Hornbeck[22] over the range 5–85° C., and by Bidwell[23] over the range − 240 to 70° C. Bidwell's results indicate a strange discontinuity between − 50° and 0° C., where the conductivity increases abnormally from 0·28 to 0·355. Thereafter the conductivity again decreases with increase in temperature and crosses Hornbeck's line which is decreasing less rapidly. Extrapolations of Hornbeck's and Bidwell's data to the melting point indicates values of 0·284 and 0·256 respectively. From the mean value for K_S of 0·27 and Porter and Simeon's figure of 1·31 for the ratio K_S/K_L, a value of 0·206 is obtained for K_L. The probable value is later considered to be at least 0·23 (*see* p. 323).

Sodium is thus another metal for which additional information is required.

Thermal Conductivity of Thallium (Melting point 302·5° C.)

The values obtained by Brown for the thermal conductivity of thallium in the solid state indicate a positive temperature coefficient and extrapolate to a value of 0·125 at the melting point :

Brown[12]

Temp., ° C.	Thermal Cond.
Solid	
45	0·0934
72	0·0938
88	0·0957
113	0·0976
125	0·0919
149	0·0927
217	0·111*
Liquid	
353	0·059*

$K_S/K_L = 2·04$

* Indicated by Brown as ' doubtful.'

The only other determination of the thermal conductivity of thallium was by Eucken and Dittrich,[24] who obtained values of 0·152 and 0·121 at − 193° and 0° C. respectively. If the thermal conductivity continues to decrease with increase of temperature according to the equation $K_S = A/T + k$ (which the authors assume holds for their results), a value of 0·114 would be reached as the melting point is approached. Thus the two extrapolated values available for K_S at 300° C. differ by about 10%, although the differences are greater near room temperature.

The value of 2·04 obtained from Brown's data for the ratio K_S/K_L agrees well with a value of 2·0 given by Schulze[25] for σ_S/σ_L.

Thermal Conductivity of Tin (Melting point 231·86° C.)

Konno's results for solid tin from room temperature upwards appear to lie reasonably well on a continuation of Lees' curve for the range − 170° to 18° C. Brown's results are roughly parallel, but nearly 10% lower in value. The results of Konno and of Brown for the liquid state are also in fair agreement. The estimated values derived from Northrup and Pratt's data are less likely to be correct and should be given but little weight. Thus, from the thermal conductivity data available, it would seem that the thermal conductivity of liquid tin is of the order of $0·07_5$ to 0·08.

On the other hand, whilst both Konno's and Brown's

Konno[1]		*Northrup and Pratt*[10]		*Brown*[12]	
Temp., °C.	Thermal Cond.	Temp., °C.	Thermal Cond.	Temp., °C.	Thermal Cond.
		Solid			
18	0·157*	202	0·187†	50	0·1428
108	0·151	219	0·135†	102	1·1404
125	0·149	222	0·120†	147	0·1390
209	0·143			160	0·1368
				209	0·1297
		Liquid			
292	0·081	285	0·057†	264	0·0815‡
417	0·079	308	0·061†	291	0·0752
498	0·078			347	0·0767
$K_S/K_L = 1·7$		$K_S/K_L = 2·0$ (?)		$K_S/K_L = 1·7$	

* Value due to Lees, assumed by Konno.

† *See* footnote (†) to bismuth values; values of $0·11_5$ to 0·11 assumed for the steel over the temperature range 160–250° C.

‡ Marked by Brown as 'doubtful.'

results agree in indicating a value of 1·7 for K_S/K_L, several observers obtain values for σ_S/σ_L in the range 2·0 to 2·2. In the light of this result it seems probable that exactly similar changes do not occur in the thermal and electrical conductivity of tin on transition from the solid to the liquid state. Further reference to this result is made later (*see* p. 322).

Thermal Conductivity of Zinc (Melting point 419·5° C.)

The published values for the thermal conductivity of zinc show closer agreement than do those for the metals considered previously. Not only are the data of Konno and Bidwell in reasonably good agreement for both the solid and liquid phases, but their results for the solid agree well with other values obtained by Bailey[4] and by Schofield.[27] It seems likely, therefore, that the thermal conductivity of liquid zinc is of the order of 0·14 and that it has a small negative temperature coefficient.

Konno[1]		*Bidwell*[26]*	
Temp., °C.	Thermal Cond.	Temp., °C.	Thermal Cond.
Solid		**Solid**	
18	0·268†	0	0·278
97	0·263	100	0·260
129	0·262	200	0·250
242	0·250	300	0·241
280	0·246	400	0·232
313	0·241		
362	0·233		
400	0·220		
Liquid		**Liquid**	
460	0·140	460	0·144
537	0·138	580	0·140
578	0·137	700	0·135
$K_S/K_L = 1·5_1$		$K_S/K_L = 1·5_6$	

* Values read from Bidwell's curve.

† Value due to Lees, assumed by Konno.

The ratio of K_S/K_L is again much less than the ratio of σ_S/σ_L—a little above 1·5 for the former ratio, compared with values ranging from 2·0 to 2·2 for the latter. This result also tends to suggest that real differences may exist in these ratios, but the matter would be much more convincing if the measurements of thermal and electrical conductivity had been carried out on the same specimen and at the same time, as this would remove any doubts as to whether the difference is associated with changes in the crystal form of the various zinc specimens.

ALLOYS FOR WHICH INFORMATION IS AVAILABLE

Thermal Conductivity of Sodium–Potassium Alloys

Two studies have been made of the thermal conductivities of sodium–potassium alloys. Hornbeck[22] measured the conductivities of solid sodium and potassium and included in this work an alloy composed of these two alkali metals having their masses proportional to their atomic weights (*i.e.*, 37% Na, 63% K). The melting point of this alloy was stated to be about 6° C. The thermal conductivity for the liquid state increases from 0·055 at 6° C. to 0·062 at 43° C. Owing to the few measurements made in the solid state, the values of K_S/K_L and σ_S/σ_L cannot be evaluated with much certainty, but it is clear that the former ratio is about 20% less than the latter.

Hornbeck[22]		*Deem and Russell*[30]	
Temp., °C.	Thermal Cond.	Temp., °C.	Thermal Cond.
Solid			
−10·6	0·0705		
− 8·9	0·0696		
Liquid		**Liquid**	
5·8	0·0547	100	0·0616
6·2	0·0551	500	0·0679
22·0	0·0584		
42·9	0·0619		
$K_S/K_L = 1·25$–$1·3$			
$\sigma_S/\sigma_L = 1·5_5$–$1·6$			

The sodium–potassium alloy studied much more recently by Deem and Russell[30] had a composition of approximately 50% by weight of each constituent. The thermal conductivity is stated to increase linearly over the range 100–500° C. It will be seen from the above data that the thermal conductivity is of the same order as that of the alloy studied by Hornbeck, but that the temperature coefficient is much smaller.

Thermal Conductivity of Four Eutectic Alloys

The following values were obtained by Brown[12] for the thermal conductivities of the eutectic alloys Pb–Bi, Sn–Pb, Sn–Zn, and Pb–Sb:

46% Pb–54% Bi, m.p. = 130° C.		*62% Sn–38% Pb, m.p. = 180° C.*	
Temp., °C.	Thermal Cond.	Temp., °C.	Thermal Cond.
Solid		**Solid**	
42	0·0236	40	0·1178
60	0·0229	70	0·1177
90	0·0230	102	0·1125
		122	0·1120
		124	0·1119
Liquid		125	0·1150
166	0·0226	139	0·1115
185	0·0229		
200	0·0233		
236	0·0244	**Liquid**	
285	0·0258	236	0·0553
		239	0·0533
$K_S/K_L = 1·04$		310	0·0610
		336	0·0615
		120	0·0710
		$K_S/K_L = 2·18$	

87% *Pb*–13% *Sb*, *m.p.* = 248° *C.*

Temp., °C.	Thermal Cond.
Solid	
34	0·0630
43	0·0637
51	0·0644
76	0·0661
77	0·0670
118	0·0678
120	0·0665
155	0·0664
168	0·0666
190	0·0650
Liquid	
316	0·0386
325	0·0396
372	0·0455
382	0·0457

$K_S/K_L = 2 \cdot 08$

92% *Sn*–8% *Zn*, *m.p.* = 200° *C.*

Temp., °C.	Thermal Cond.
Solid	
35	0·1425
64	0·1469
91	0·1474
115	0·1497
120	0·1440
148	0·1505
Liquid	
215	0·0570
343	0·0731
433	0·0878

$K_S/K_L = 3 \cdot 05$

Electrical conductivities were not measured, and no other data have been noticed with which these results can be compared. It is interesting to note that the mixture of roughly equal amounts of lead and bismuth, which individually behave in opposite ways, has only a very small discontinuity in thermal conductivity on changing from solid to liquid. The eutectic alloy of tin and zinc, on the other hand, has an unusually high value for the ratio K_S/K_L.

Again, there is need for much more experimental work on the thermal conductivity of molten alloys.

COMPARISON BETWEEN THEORY AND EXPERIMENT FOR CONDUCTIVITY CHANGES ON MELTING

To facilitate reference the available data for K_S/K_L and σ_S/σ_L have been brought together in Table I. With regard to the latter ratio, the columns headed 1926 and 1927 contain the results of summaries of existing data published in those years by Perlitz[28] and by Matuyama,[7] respectively. The column headed Final Mean contains values obtained by taking the arithmetic average of these results and of any subsequent determinations.

Bridgman[31] pointed out that the electric resistance of metallic elements changes abruptly on fusion in such a way that the phase with the smaller specific volume has the smaller specific resistance. Perlitz[28] considered that there should be some correlation between the change in specific resistance on fusion and the crystal form of the solid, since the atoms might be regarded as being randomly distributed in the liquid state. His tabulation of the mean values of σ_S/σ_L showed that, apart from mercury, the metals could be divided into three groups : Bi, Sb, and Ga, for which σ_S/σ_L was approximately $\frac{1}{2}$*; Na, Li, K, Rb, Cs, and Al, for which σ_S/σ_L was approximately $\frac{3}{2}$; and Ag, Cd, Pb, Tl, Cu, Zn, Te, Sn, and Au, for which σ_S/σ_L was approximately 2. The crystal lattices of Bi and Sb were of the same rhombohedral hexagonal type, whilst that of Ga had not been determined. The alkali metals of the second group form body-centred cubic lattices, but aluminium is face-centred cubic. Most of the metals of the third group have face-centred cubic or hexagonal lattices which belong to the close-packed structure. Thus, there seemed to be a reasonable amount of support for Perlitz's conclusion, and the subsequent determination of σ_S/σ_L for aluminium has tended to move aluminium into the close-packed group to which it belongs and for which the ratio is about 2.

Mott[32] has obtained a theoretical expression for the change in electrical resistivity on melting. It is assumed that the atoms in a liquid vibrate about slowly varying mean positions with a frequency γ_L, and the ratio of γ_L to γ_S, the atomic frequency of the solid, is calculated from the observed latent heat L kilo-joules/g. atom, and the melting point T_M ° K. For normal metals, the change in electrical conductivity on melting is related to the change in atomic frequency to a first approximation by :

$$\sigma_S/\sigma_L = (\gamma_S/\gamma_L)^2 = e^{80L/T_M} \quad \ldots\ldots\ldots\ldots (1)$$

Values calculated in this way are included in the last column of Table I.

For bismuth and antimony, σ_S is less than σ_L, and the ratio has not the high values suggested by the

* The present author has recently found σ_S/σ_L to be about $\frac{5}{4}$ for Ga, which therefore disagrees with the above generalizations of Bridgman.

Table I

COLLECTED RESULTS FOR THE RATIOS K_S/K_L AND σ_S/σ_L

	K_S/K_L								σ_S/σ_L							
									1926		1927		Subsequent Observers			
	Konno	Northrup and Pratt	Brown	Bidwell	Gehlhoff and Neumeier	Porter and Simeon	Mean	Calculated Value (Rao)	No.	Mean	No.	Mean	Matuyama	Pietenpol and Miley[29]	Final Mean	Calculated Value (Mott)
Al	1·47	..	..	..	..	..	1·47	1·8	1	1·65	2	1·63	2·20		1·82	2·0‡
Sb	1·1	..	..	..	..	..	1·1		2	0·69	2	0·71	0·717		0·71	5·6
Bi	0·45	0·65	..	..	..	..	0·55		8	0·464	5	0·486	0·486	0·44*	0·47	5·0
Cd	..	..	2·4	..	..	..	2·4		3	1·91	3	1·91	1·83		1·89	2·3
Pb	1·83	..	..	1·24	..	..	1·53	1·88	3	1·97	3	1·97	2·055	1·93*	1·98	1·87
Hg	..	..	..	..	2·95	3·91	3·43		6	4·23	3	2·92	2·515		3·66	2·23
Na	..	..	..	..	..	1·31	1·31	1·58	6	1·45	3	1·47	..		1·46	1·7†‡
Tl	..	..	2·04	..	..	..	2·04		1	2·0	..	..	..		2·0	2·3
Sn	1·7	2·0	1·7	..	..	..	1·8	2·81	5	2·13	4	2·13	2·095	1·98*	2·11	(3)†
Zn	1·51	..	..	1·56	..	..	1·53	2·23	2	2·05	3	2·10	2·148	2·26*	2·11	2·3

* Derived from a plot of Pietenpol and Miley's results and by extrapolation of the linear portion of the curve for the solid state.

† Mott suggests that L is not known accurately for Sn, an alternative explanation may be that T_M is less than 2Ⓗ where Ⓗ is the characteristic temperature (260 for Sn).

‡ Another case where T_M is relatively large compared with Ⓗ; by use of alternative expressions which partially allow for this, Mott obtains values of 1·8 and 1·58 for Al and Na respectively.

theory, so it is concluded that the effective number of free electrons increases by a factor of about 10 when these metals melt. Mercury, on the other hand, is the only metal in which the increase of resistance on melting is considerably greater than that predicted by the theory. The agreement obtained between theory and experiment for the normal metals is rather surprising and tends to confirm that the extra resistance in the liquid phase is due to the greater amplitude of the atomic oscillations and not, to any large extent, to the atoms being more irregularly arranged in the liquid than in the solid. Mott's treatment suggests that the electrical resistivity of the liquid metal should increase with the temperature; the temperature coefficient being inversely proportional to the absolute temperature.

Rao[33] has since published an almost identical treatment for the ratio K_S/K_L. Bidwell's experiments[13] are accepted by Rao as showing that the thermal conductivity of a molten metal is almost entirely due to electrons. If this is the case, it follows immediately that the ratio of the thermal to the electrical conductivity of a molten metal is constant and, just as indicated above, to a first approximation, the ratio :

$$K_S/K_L = (\gamma_S/\gamma_L)^2 = e^{80L/T_M} \quad \ldots\ldots\ldots\ldots (2)$$

The calculated values given by Rao are included in Table I. Where his values differ from Mott's, Rao has presumably used the more exact expression for the ratio (γ_S/γ_L), which Mott showed to be necessary when the melting point is not several times greater than the characteristic temperature.

CONSIDERATION OF BIDWELL'S INTERCEPT LAW

The work of Bidwell, which was mentioned in the previous section, will now be considered in some detail as it has a direct bearing on the subject of the thermal conductivity of molten metals.

In 1928, Bidwell[34] claimed that data he had obtained for the thermal conductivity K and the atomic heat AC of lithium, sodium, zinc, and lead, over the temperature range $-250°$ to $+200°$ C., conformed to the relation :

$$K/AC = k_1/T + k_2 \quad \ldots\ldots\ldots\ldots (3)$$

where T is the absolute temperature and k_1 and k_2 are constants. Data taken from the Landolt-Börnstein tables for copper, aluminium, and mercury, also gave straight lines when K/AC, was plotted against $1/T$.

When presenting these results, Bidwell stated : " It is possible that the second term on the right, that is the intercept, is the value of K/AC, which has to do with the transport of heat by atomic collisions. According to the experimental law this term is independent of temperature. The behaviour of the mercury line suggests that this has approximately the same value as for the liquid metal."

For mercury, the value of the atomic portion of the thermal conductivity indicated by this intercept is 0·0335. This agrees with Gehlhoff and Neumeier's value for the thermal conductivity of mercury at 70° C., but is greater than all the other values at normal temperatures. The intercept actually corresponds to $1/T = 0$, that is, to an infinite value of T, so there may be some grounds for the higher value since liquid mercury appears to have a positive coefficient of thermal conductivity.

The subject received further consideration in a paper by Bidwell,[13] published in 1940, where the original equation is modified to :

$$K/\delta AC = k_1'/T + k_2' \quad \ldots\ldots\ldots\ldots (4)$$

by division of the left-hand side by δ, the density, and is given the following interpretation in the light of modern theory. Wilson[35] has shown that the thermal conductivity can be resolved into two parts and written as :

$$K = k_g + k_e \quad \ldots\ldots\ldots\ldots (5)$$

where k_g is the conductivity due to the lattice, and k_e that due to electrons. At high temperatures, k_e becomes constant, and Wilson shows that the two terms can be represented by :

$$K = \tfrac{1}{3}\delta ACLu + \frac{\pi^2}{3} n \frac{\tau R^2 T}{m} \quad \ldots\ldots\ldots\ldots (6)$$

where L = the mean free path in the lattice
u = the velocity of sound in the lattice
n = the number of free electrons per unit volume
R = the molecular gas constant
m = the mass of the electron
and τ = the relaxation time, the average time between collisions.

Bidwell divides throughout by δAC, obtaining

$$K/\delta AC = \tfrac{1}{3}Lu + \frac{\pi^2}{3} \frac{n\tau R^2 T}{m\delta AC} \quad \ldots\ldots\ldots\ldots (7)$$

and considers that this must be equivalent to his own equation (4), as experimental data reported in his paper for the three metals lead, tin, and zinc, yield straight lines if $K/\delta AC$ is plotted against $1/T$. This conclusion regarding the equivalence of the equations implies that Lu varies as T^{-1}, and that τ varies as T^{-1}, since the product δC is assumed to be approximately independent of temperature.

In his 1940 paper, Bidwell states quite clearly that the results for tin, lead, and zinc are in agreement with the suggested law, the first term on the right relating to conduction by the crystal lattice, and the second term to conduction by electrons ; it is further stated that the intercepts give the value of K/AC for the molten state. Thus, if this law should prove to be generally true, estimated values for the thermal conductivities of liquid metals could be obtained by plotting the thermal data for the solid state in the manner suggested by Bidwell and extrapolating to the ordinate $1/T = 0$. In view of the importance of this result it is appropriate to re-examine the evidence supplied by Bidwell.

The following values for the elastic and non-elastic portions of thermal conductivity at 0° C., are taken from Bidwell's paper[34] (*loc. cit.*, p. 314, Table II) of 1928 :

Metal	Elastic Wave Portion $K_1 = k_1 AC/T$	Atomic Portion $K_2 = k_2 AC$
Lithium	0·150	0·006
Sodium	0·092 (−10° C.)	0·169
	0·124 (+10° C.)	
Zinc (polycrystal)	0·084	0·192
Zinc (single crystal)	0·098	0·192
Aluminium	0·127	0·358
Copper	0·920	0·171
Lead	0·018	0·068
Mercury	0·0308 (solid extrapolated)	0·0335

The table includes each of the metals for which Bidwell had obtained a straight line when K/AC was plotted against $1/T$. The elastic wave portion of the conductivity is derived from the slope of this line, and the non-elastic or atomic portion from the intercept. Since the two samples of zinc give the same intercept, it is concluded that the increase in thermal conductivity of the single crystal occurs entirely in the elastic wave component. Similarly, in the case of sodium, which was found to suffer an abrupt change in thermal conductivity just below 0° C., the increased conductivity is entirely in the elastic wave portion, since the slope is increased but the intercept is unchanged. It is only with regard to mercury that the suggestion is made that the atomic portion is approximately the thermal conductivity of the molten metal, and the paper concludes with the statement : " Data on other metals are being obtained to test the equation and theory further, particularly, for the liquid state."

The 1940 paper[13] gives results for the three metals, tin, lead, and zinc, in the form of graphs of K against T and of $K/AC\delta$ against $1/T$.

The data presented for tin are Konno's (*see* p. 319), Bidwell having made no measurements on tin. The intercept, as measured on the published curve, is 0·00152, which, assuming values of 7·0 for the atomic heat of molten tin and of 7·0 for its density, gives a value of 0·0745 for the thermal conductivity. This value is certainly in agreement with the data given on p. 319 for the thermal conductivity of liquid tin, particularly at high temperatures.

In the case of lead, Konno's data for the liquid state are not plotted, as Bidwell considered Konno's values to be far too low for $K/AC\delta$ to agree with the intercept. Instead, Bidwell's own newly determined values are used. The value of the conductivity component calculated from the intercept, assuming $AC = 6{\cdot}8$ and $\delta = 10{\cdot}7$, is 0·058. This is less than the value of 0·068 given in Bidwell's earlier paper (*see* p. 321) and is, as Bidwell says, in agreement with his own results for the thermal conductivity of molten lead, but is 50% greater than Konno's values (*see* p. 317). Thus the results for which the law holds appear to have been selected. Konno's data for lead which happen to disagree are omitted, whilst Konno's data for tin which agree are accepted without question and are quoted in support of the theory.

The data plotted for zinc are somewhat puzzling. Bidwell has extended his earlier data on polycrystalline zinc to include the molten state, and he gives both sets of results. They are in good agreement for the solid state over their common range, so that it is to be anticipated that the value previously obtained from the intercept and given on p. 321 should still hold. This value is 0·192, which is very different from the values of 0·144 to 0·135, obtained for molten zinc in the range 460–700° C. (*see* p. 319).

Measurement of the intercept in his plot of $K/AC\delta$ against $1/T$ indicates a value of 0·0043. The atomic heat of molten zinc is 7·2 (*see* the Appendix), and the density is 6·6 (International Critical Tables). Hence the value derived for the thermal conductivity is $0{\cdot}0043 \times 7{\cdot}2 \times 6{\cdot}6 = 0{\cdot}205$. This value is comparable with that indicated by Bidwell's earlier results, but is no less than 50% greater than his value for molten zinc—with which he claims that agreement is obtained.

Incidentally, it was for zinc that fairly good agreement was noted between the results of Konno and Bidwell for both solid and liquid phases, so that this would seem to be an instance in which the intercept quite definitely does not agree with the thermal conductivity of the molten state.

In the case of aluminium, the evidence is again inconclusive as the two sets of data available for the thermal conductivity in the liquid state differ appreciably. Bidwell and Hogan's claim, that their recent results for this metal are in agreement with the intercept theory, has been questioned elsewhere.[36]

To sum up, since, at the moment, tin appears to be the only metal for which the intercept agrees closely with the thermal conductivity of the molten state, it is clear that this suggested method cannot be accepted as a means for estimating the thermal conductivity of molten metals. There is clearly need for further experimental work in this field before any such relationship as that suggested by Bidwell can be accepted.

CONSIDERATION OF THE LORENZ FUNCTION OF MOLTEN METALS

A more commonly accepted method for the estimation of thermal conductivities is to measure the electrical resistivity, since, according to the Wiedemann-Franz-Lorenz Law, the ratio $K/\sigma T$ should be constant. In general, this ratio, L, termed the Lorenz function, undergoes appreciable changes in value at low temperatures, but tends to become more nearly constant at high temperatures. This tendency is understandable since the law applies theoretically to conduction by means of electrons. In fact, if, as Bidwell stated, his intercept gives the electronic conduction term and this had been the conductivity of the molten state, it would have been reasonable to suppose that the Lorenz function of a liquid metal would have its theoretical value of $0{\cdot}58 \times 10^{-8}$. A similar assumption was implied by Rao when he adapted Mott's theoretical treatment of the ratio σ_S/σ_L to apply to K_S/K_L. There is, however, the further assumption that in K_S the electronic component is already predominant. This is highly probable since, for most normal metals, L approximates to the theoretical value before the melting point is reached.

The following values of the Lorenz function of molten metals, obtained from existing data, should be accepted with some reserve, as the two conductivity measurements are usually due to different observers, as is indicated by initial letters of surnames :

Lorenz Function of Molten Metals

Property at :	M.p., T_M	$(T_M + 100)$	$(T_M + 200)$	Observer
Aluminium				
K	0·226	0·215	0·209	Konno
	0·124	0·148	0·17	Bidwell and Hogan
$10^6\rho$	20·4	21·7	23·2	Tsutsumi
	25·7	26·8	28	Matuyama
10^8L	0·49; 0·27	0·45; 0·31	0·43; 0·35	K.T.; B.&H.T.
	0·62; 0·34	0·56; 0·38	0·52; 0·42	K.M.; B.&H.M.

Property at:	M.p., T_M	$(T_M + 100)$	$(T_M + 200)$	Observer
Antimony				
K	0·052	0·05		Konno
$10^6\rho$	110	114		Tsutsumi
	115	116·2		Matuyama
10^8L	0·63	0·55		K.T.
	0·66	0·56		K.M.
Bismuth				
K	0·04	0·038	0·037	Konno
$10^6\rho$	128	129	130	Tsutsumi
10^8L	0·94	0·76	0·65	K.T.
Cadmium				
K	0·108	0·117		Brown
$10^6\rho$	32·8	33·0		Matuyama
10^8L	0·60	0·56		B.M.
Lead				
K	0·0394	0·0384		Konno
	0·0587	0·059		Bidwell
$10^6\rho$	96·4	103		Mean value
10^8L	0·63	0·57		K.MV
	0·94	0·87		B.MV
Mercury				
K	0·0218	0·032	0·04	Gehlhoff and Neumeier
$10^6\rho$	90·9	99·5	110	Mean value
10^8L	0·85	0·95	1·01	G. & N.MV
Mean values at 20° C.: $K = 0·02$, $10^6\rho = 95·76$, $10^8L = 0·65$				
Tin				
K	0·082	0·0804	$0·078_5$	Konno
	0·075	0·076		Brown
$10^6\rho$	47·25	51·15	55·22	Pietenpol and Miley
10^8L	0·77	0·68	$0·61_5$	K.P. & M.
	0·70	0·64		B.P. & M.
Zinc				
K	0·141	0·139	0·137	Konno
	0·147	0·143	0·139	Bidwell
$10^6\rho$	37	37	38	Tsutsumi
10^8L	0·75	0·65	0·58	KT
	0·78	0·67	0·59	BT
Na–K Alloy (masses propl. at.wt.)				
K	$0·05_5$			Hornbeck
$10^6\rho$	34			Hornbeck
10^8L	0·67			H.H.

These values provoke the following comments: Bidwell and Hogan's values for the thermal conductivity of molten aluminium would seem to be much too low, since it seems unlikely that the Lorenz function of a molten metal should be less than the theoretical value. Konno's values for this metal may also be low, particularly at high temperatures.

The high values of the Lorenz functions of zinc and tin immediately above the melting point are of interest. For each metal, similar results are obtained from two independent sets of data, and the high values suggest that the thermal conductivity of a molten metal may not be predominantly of an electronic nature. This applies just above the melting point, but there is seen to be a tendency to approach the theoretical value at higher temperatures. This might tend to justify Bidwell's identification of the intercept term as the electronic contribution to the thermal conductivity. On the other hand, Rao's application of Bidwell's results to the region near the melting point would not be justified.

The marked rise in the Lorenz function of mercury is unusual and would suggest that Gehlhoff and Neumeier's values for the thermal conductivity may be in error.

The other results, apart from those for bismuth, are fairly normal. Only one molten alloy is included, but it is satisfactory that in this instance both conductivities were measured by the same observer. The Lorenz function of this Na–K alloy also increases on melting, and shows a further increase to a value of 0·71 at 43° C.—some 36° C. above the melting point.

The general indications are that just above the melting point the Lorenz function of most metals may have values ranging from $0·6 \times 10^{-8}$ to $0·8 \times 10^{-8}$, but that at some 200° C. above the melting point the Lorenz function might be expected to exceed the theoretical value by not more than about 10%.

On the basis of these generalizations, the following estimated values of thermal conductivities for certain other molten metals with low melting points, for which experimental data are not yet available, have been calculated:

	M.p., T_M, °C.	Elect. Resist. ($10^6\rho$) (T_M)	(T_M+200)	Thermal Cond.* (T_M)	(T_M+200)
Cs	28·45	36·0	...	0·05 to 0·065	...
Ga	29·78	25·9	...	0·07 to 0·09	...
Rb	39·0	22·5	...	0·085 to 0·11	...
K	63·6	13·6	26	0·15 to 0·20	0·13
Na	97·6	$9·6_5$	16·7	0·23 to 0·3	0·20
In	156·4	29	...	0·09 to 0·12	...
Li	186	31·7	...	0·085 to 0·11	...

* Cal. cm.$^{-1}$ sec.$^{-1}$ °C.$^{-1}$

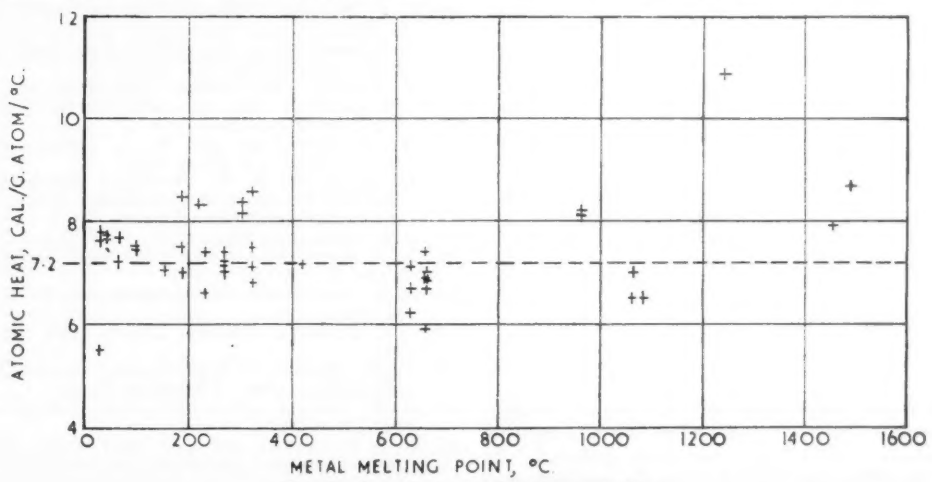

Fig. 2—Atomic heat of molten metals

The alloys are too numerous for them to be treated in a similar manner. Reference should, however, be made to the potassium–rubidium series which the results of Kurnakow and Nikitinski[37] show has a lower electrical resistivity than other alloy systems mentioned in the literature. The alloy 40% Rb–60% K, on a weight basis, is molten at 50° C. and has a resistivity of $16{\cdot}9 \times 10^{-6}$ ohm cm. According to the foregoing conclusions, the thermal conductivity of this alloy would be expected to be $0{\cdot}13 \pm 0{\cdot}02$ cal. cm.$^{-1}$ sec.$^{-1}$ C.$^{-1}$

Acknowledgment

This review has been made as part of the work of the National Physical Laboratory and is published by permission of the Director of the Laboratory.

Appendix

ATOMIC HEAT OF MOLTEN METALS

By I. Backhurst, M.Sc.

In the following table and in Fig. 2 are experimental values for the atomic heat of molten metals, taken from the International Critical Tables (I.C.T.) and the tables of Landolt-Börnstein :

Atomic Heats of Molten Metals

	M.p., ° C.	Atomic Heat, cal./g. atom/° C. I.C.T.	Landolt-Börnstein
Cs	28·45	7·65 at 50° C.	
Ga	29·78	5·5 at 119° C.	6·81±0·10 at 21–100° C.
Rb	39·0	7·65 at 50° C.	7·85 at 39·0–100° C. to 5%
K	63·6	7·2 at m.p.	7·7 at m.p. to 100° C. to 5%
Na	97·6	7·65 at 100° C.	7·50 at m.p. to 178° C. to 2%
In	156·4		7·04 at m.p. to 184·2° C.
Li	186		8·5 at m.p. to 427° C. 7·5 at 427–727° C. 7·0 at 727–1227° C.
Se	220		8·35 at m.p. to 300° C. to 3%
Sn	231·86	7·4	6·6 at m.p. to 1000° C.
Bi	269	7·4 at 400° C.	$7{\cdot}20_8$ at 272·9° C. $7{\cdot}13_9$ at 304·2° C. $6{\cdot}98_3$ at 370·6° C.
Tl	302·5		8·18 at 400° C. 8·40 at 500° C.
Cd	320·9	8·6 at m.p.	7·50 at m.p. to 720° C. 7·13 constant
Pb	327·4		6·8 at m.p. to 1000° C. to 5%
Zn	419·5		7·18 at m.p. to 865° C.
Sb	630·5	6·7 at m.p.	7·15 at m.p. to 1000° C. to 5% $6{\cdot}198 + 7{\cdot}21 \times 10^{-4}t°$
Mg	649		7·4 to 10% $6{\cdot}91 + 1{\cdot}3 \times 10^{-3}\,(t° - 651)$
Al	660	6·7 at m.p.	7·0 at m.p. to 1000° C. to 3% $6{\cdot}84 + 1{\cdot}26 \times 10^{-3}\,(t° - 727)$ $5{\cdot}90 + 1{\cdot}295 \times 10^{-3}\,t°$
Ag	960·5	8·1 at m.p. to 1100° C.	8·2 at m.p. to 1300° C. to 3%
Au	1062·8	6·5 at 1100° C.	7·0 at m.p. to 1300° C. to 5%
Cu	1083	6·5 at m.p.	
Mn	1242		10·88
Ni	1455	7·9 at m.p.	
Co	1490		8·68 at m.p. to 1600° C.

A few values which appeared to be much too high or too low have been omitted, and metals with melting points above 1500° C. have not been included. In view of the difficulties attendant on the measurement of atomic heat of molten metals, it is probable that the accuracy attained is insufficient to justify much more than a general statement that, within the range considered, the atomic heat of any molten metal within a few hundred degrees above its melting point is about 7 cal./g. atom/° C.

References

1. S. Konno : *Science Reports of the Tohoku Imperial University*, 1919, vol. 8, p. 169.
2. C. C. Bidwell and C. L. Hogan : *Journal of Applied Physics*, 1947, vol. 18, p. 776.
3. C. H. Lees : *Philosophical Transactions of the Royal Society*, 1908, series A, vol. 208, p. 381.
4. L. C. Bailey : *Proceedings of the Royal Society*, 1931, series A, vol. 134, p. 57.
5. E. F. Northrup : *Transactions of the American Electrochemical Society*, 1914, vol. 25, p. 373.
6. H. Tsutsumi : *Science Reports of the Tohoku Imperial University*, 1918, vol. 7, p. 93.
7. Y. Matuyama : *Ibid.*, 1927, vol. 16, p. 447.
8. L. Lorenz : *Annalen der Physik und Chemie*, 1881, vol. 13, pp. 422, 582.
9. L. de la Rive : *Archives des Sciences Physiques et Naturelles*, 1863, vol. 18, p. 362.
10. E. F. Northrup and F. R. Pratt : *Journal of the Franklin Institute*, 1917, vol. 184, p. 675.
11. W. Jaeger and H. Diesselhorst : *Wissenschaftliche Abhandlungen der physikalisch-technischen Reichsanstalt*, 1900, vol. 3, p. 269.
12. W. B. Brown : *Physical Review*, 1923, vol. 22, p. 171.
13. C. C. Bidwell : *Ibid.*, 1940, vol. 58, p. 561.
14. G. Gehlhoff and F. Neumeier : *Verhandlungen der deutschen physikalischen Gesellschaft*, 1919, vol. 21, p. 201.
15. A. W. Porter and F. Simeon : *Proceedings of the Physical Society*, 1915, vol. 27, p. 307.
16. H. F. Weber : *Annalen der Physik und Chemie*, 1880, vol. 10, p. 472.
17. A. J. Ångström : *Ibid.*, 1864, vol. 123, p. 628.
18. A. Berget : *Comptes Rendus*, 1887, vol. 105, p. 224.
19. R. Weber : *Annalen der Physik*, 1903, vol. 11, p.1047.
20. H. R. Nettleton: *Proceedings of the Physical Society*, 1915, vol. 27, p. 129.
21. H. Reddemann : *Annalen der Physik*, 1932, vol. 14, p. 139.
22. J. W. Hornbeck : *Physical Review*, 1913, vol. 2, p. 217.
23. C. C. Bidwell : *Ibid.*, 1926, vol. 28, p. 584.
24. A. Eucken and K. Dittrich : *Zeitschrift für physikalische Chemie*, 1927, vol. 125, p. 211.
25. A. Schulze : *Zeitschrift für Metallkunde*, 1923, vol. 15, p. 155 ; 1924, vol. 16, p. 48.
26. C. C. Bidwell : *Physical Review*, 1939, vol. 56, p. 594.
27. F. H. Schofield : *Proceedings of the Royal Society*, 1925, series A, vol. 107, p. 207.
28. H. Perlitz : *The London, Edinburgh, and Dublin Philosophical Magazine and Journal of Science*, 1926, vol. 2, p. 1148.
29. W. B. Pietenpol and H. A. Miley : *Physical Review*, 1929, vol. 34, p. 1588.
30. H. Deem and H. W. Russell : Manhatten Declassified Document 342, Oakridge, Tenn., U.S.A., 1946 : Document Sales Agency.
31. P. W. Bridgman: *Physical Review*, 1921, vol. 17, p161.
32. N. F. Mott : *Proceedings of the Royal Society*, 1934, series A, vol. 146, p. 465.
33. M. R. Rao : *Indian Journal of Physics*, 1942, vol. 16, p. 155.
34. C. C. Bidwell : *Physical Review*, 1928, vol. 32, p. 311.
35. A. H. Wilson : "Semi Conductors and Metals." Cambridge, 1939 : University Press.
36. R. W. Powell : *Journal of Applied Physics*, 1948, vol. 19, p. 995.
37. N. S. Kurnakow and A. J. Nikitinski : *Zeitschrift für anorganische und allgemeine Chemie*, 1914, vol. 88, p. 151.

Constitution of Iron–Nickel–Chromium Alloys at 650° to 800° C.

By W. P. Rees, M.Sc., B. D. Burns, M.Sc., and A. J. Cook

SYNOPSIS

A considerable number of iron–nickel–chromium alloys were prepared with a range of compositions extending from 40 to 80% of iron, 0 to 60% of nickel, and 0 to 50% of chromium, using component elements of very high purity. Metallographic and X-ray investigations of the alloy structures were made to determine the limits of composition at which the sigma constituent did not exist as an equilibrium component after annealing the alloys at 650° and 800° C. The results show that these limits extend into a range of compositions considerably lower in chromium content than has been previously stated.

A feature of the alloys was the sluggish behaviour of the more highly alloyed compositions during annealing treatments, and it was found that the transition from the γ phase to the $(\gamma+\sigma)$ phase was not always a single-stage reaction, but could be preceded by the precipitation of an intermediate α phase which in some compositions was very persistent.

Introduction

THE researches of Schafmeister and Ergang[1] in 1939 and of Bradley and Goldschmidt[2] in 1941 established the nature of the phases occurring in the iron–nickel–chromium system and the general outline of the phase diagram, but there remained significant discrepancies in the definition of the boundaries. In view of the increasing importance of the applications of nickel–chromium steels at high temperatures, the Alloy Steels Committee of the British Iron and Steel Research Association set in hand, at the National Physical Laboratory, a re-investigation of the system, with special reference to the constitution at 550° to 800° C. of alloys in the range of composition covered by the common heat-resisting steels.

The work has proved tedious because many of the alloys have undergone very slow changes in this range of temperature. In many cases it is still impossible to say what the constitution will be when a state of equilibrium has ultimately been reached, and this report describes essentially the conditions that are established when the alloys are annealed for periods up to 100 or 200 days at 650° C. or 800° C. Some annealings at 550° C. have been made, but the changes at this temperature are even slower than at 650° C., and a useful account of the behaviour at 550° C. cannot yet be given.

At this stage, the work has been carried out on the purest possible alloys. Commercial steels contain proportions of carbon, silicon, manganese, and other constituents, which are known to have a significant effect on their constitutional behaviour, but it has been considered that the more complex steels cannot be understood until the behaviour of the simple iron–nickel–chromium alloys has been established. The changes of constitution which occur slowly in many of the alloys are no doubt accompanied by changes of their mechanical properties, but until the mechanical properties have been examined it will be impossible to say whether the changes are favourable or unfavourable.

The compositions of the alloys fall within the range 0 to 60% of nickel and 0 to 50% of chromium.

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Mr. Rees, Mr. Burns, and Mr. Cook are at the Metallurgy Division of the National Physical Laboratory, Teddington, Middlesex.

EXPERIMENTAL PROCEDURE

Preparation of the Alloys

All the alloys have been made from pure iron and pure chromium prepared in the Metallurgy Division of the National Physical Laboratory, and from pure nickel powder supplied by The Mond Nickel Co., Ltd. At first the three metals were independently freed from oxygen by treatment at high temperatures with dry, purified hydrogen, the iron and nickel being melted in hydrogen, the chromium being treated at a temperature of about 1500° C., and the alloys were made by melting the deoxidized metals *in vacuo*. Later it was found that the alloys could be prepared free from oxygen by melting together the three metals under a hydrogen pressure of about 16 cm. of mercury, without prior deoxidizing treatments of each individual metal. The ingots were made in pure alumina pots coated on the inside with a thin wash of thoria. Melting was carried out in a valve-oscillator high-frequency furnace, the input to which could be controlled to within close limits. This was important as the fairly lengthy treatments of the molten alloys in hydrogen must be carried out at as low a temperature as possible in order to minimize attack on the crucible material. Attack of the crucible is especially liable to occur with alloys of high chromium content.

The melts, generally of 100 g. weight, were allowed to cool in the crucible *in vacuo*, the hydrogen being removed before solidification. The resulting ingots were polished, and examined under the microscope to make certain that they were reasonably free from small inclusions of oxide before any further work was carried out on them.

In making the alloys to a predetermined composition, allowance had to be made for a different

Table I
CHEMICAL ANALYSIS OF ALLOYS

Alloy	Ni, %	Cr, %	Fe, %	Other Elements, %
8F18	8·12	18·31	ND*	
8F20	8·09	20·22	ND	
9F20	8·64	19·86	71·30	N_2 0·008
11F22	10·95	22·83	66·20	
11F40	11·25	40·40	48·37	Insol. 0·07
12F20†	11·27	21·18	67·45	Insol. 0·05, Si Nil
12F19	12·84	18·40	ND	
13F18	13·36	17·20	69·23	
13F45	13·30	48·52	38·16	
14F15†	14·47	15·19	70·25	Insol. 0·03
14F20	14·26	18·96	ND	C 0·004
14F28	15·30	27·82	56·82	Insol. 0·07, Si 0·002
16F11	16·02	10·76	72·94	
17F23	16·72	22·91	60·20	
19F30	19·10	33·20	48·30	
18F5†	17·99	5·03	76·94	
19F20†	19·51	19·36	61·01	
20F25	20·16	24·65	ND	C 0·006
21F17	21·07	16·41	62·65	
23F14†	23·34	14·71	62·01	
23F25†	23·90	27·00	49·10	Insol. 0·05, C 0·0045
26F24	26·33	22·83	50·75	
26F26	25·13	26·07	48·72	Insol. 0·07
26F31	26·43	31·24	42·19	
27F37	27·27	39·05	33·62	Insol. 0·06
30F19	29·87	18·93	ND	
30F23	30·01	23·00	ND	
31F26	31·25	25·78	42·77	N_2 0·004, Si 0·006
32F14	30·78	13·97	ND	
33F33†	34·01	33·24	32·69	Insol. 0·06
35F12	36·90	12·40	50·90	
35F22	35·29	22·05	42·42	N_2 0·004, Si 0·004
37F29	36·47	29·68	ND	
38F18	38·34	18·22	ND	
41F16	40·88	16·75	42·21	
41F24	41·58	22·40	ND	
42F23	42·29	23·00	ND	C 0·002
45F19	44·86	19·96	ND	
48F22	50·90	20·50	28·60	Insol. 0·07
50F30	50·04	30·07	ND	
0F30†	...	28·70	ND	
0F32	...	32·46	ND	
0F38	...	38·72	ND	
1F20	0·95	19·82	79·06	
1F25	1·10	26·34	72·42	
1F30	1·16	30·72	67·82	
2F20	2·24	22·48	74·98	
2F25	2·02	24·48	73·38	
2F28	2·07	27·08	70·63	
3F18	3·03	17·36	79·40	
3F20	3·07	19·18	77·65	
3F25	3·00	24·56	72·27	
4F25	4·33	25·00	ND	
5F18	5·10	17·80	ND	
5F20	5·08	20·40	ND	
5F22	4·84	21·80	ND	
5F30	4·99	29·68	ND	
5F34	5·45	33·63	ND	
5F38	4·98	36·97	58·24	
6F30	6·40	30·10	ND	
6F35†	6·73	34·94	58·09	Insol. 0·05, Si Nil
9F37†	10·13	38·27	51·42	
20F0†	20·43	...	79·61	
25F7	25·15	7·24	ND	
31F0†	33·94	Nil	66·12	
37F7	37·05	7·59	ND	
35F25	34·70	25·80	ND	C 0·003
60F10	59·42	10·01	ND	
70F30	71·00	29·00	Nil	
80F20	80·50	19·50	Nil	
90F10	89·80	10·20	Nil	

* Not determined.
† Alloys made by melting previously deoxidized metals *in vacuo*; all other alloys were made by melting in dry purified hydrogen.

percentage loss of each of the metals by evaporation. If, for any reason, the alloys have to be maintained under vacuum conditions near the melting point for any considerable time, the loss by evaporation will be very high—up to 5% of the weight of the charge. The method of preparation of each alloy is indicated in Table I.

Chemical Analysis

The nominal compositions and the compositions determined by chemical analysis are given in Table I. The compositions are believed to be correct to within $\pm 0{\cdot}2\%$. The samples for analysis were obtained by machining over the whole cross-section of the ingot after it had been heated for three days at 1350° C. The nominal composition is used in the text to identify the alloys, the first number giving the nickel and the second number the chromium percentage by weight.

Heat-Treatments

All heat-treatments of ingots or of specimens cut from them have been carried out either *in vacuo* or in hydrogen. This was necessary, especially in the case of the high-chromium alloys, to avoid contamination by nitrogen, and to a lesser extent by oxygen.

To eliminate coring, as far as practicable, all the ingots were heated and maintained for three days at 1350° C. in a platinum-wound furnace, evacuated to a pressure of 0·02–0·03 mm. of mercury. After this homogenizing treatment they were heated in hydrogen to 1000° C., or higher, for 1 hr. and quenched in cold water. This latter treatment rendered the majority of the alloys either wholly austenitic or ferritic. Reference is made later to the desirability of starting with the alloys in a single-phase condition if equilibrium is to be attained in a reasonable time at lower temperature. These preliminary treatments resulted in a fairly large grain size in the majority of the alloys.

After quenching from 1000° C., small lump samples —usually quadrants cut from a thin transverse slice through the centre of the ingot—were taken from each alloy for heat-treatment at 550°, 650°, and 800° C. Powder samples for X-ray examination were obtained by filing over the whole longitudinal section of the ingots, or by sawing with a jeweller's fine-toothed saw. Both lump and powder samples were sealed into evacuated silica tubes for heat-treatment in thermostatically controlled furnaces, the temperatures of which were maintained constant to within $\pm 2°$ C. The samples were quenched from the annealing temperatures by plunging the silica tubes into water and crushing those containing the lump samples.

Determination of the Presence of Sigma Phase

Microscopic examination of the lump samples and X-ray examination of powder samples have been used to establish the phases present in the alloys. In a number of cases the identity of phases visible under the microscope has been confirmed by X-ray examination of the lump samples, and in other cases parts of the powder samples used for X-ray examination have been mounted in bakelite and examined under the microscope. In addition, small pieces of the alloy quenched from a high temperature have been cold

rolled to thin strip and subsequently annealed at 650° C. and 800° C. This has greatly accelerated the approach to equilibrium, and because of the resulting fine grain size a satisfactory comparison of the X-ray examination, using the glancing-angle technique, with the microscopic examination of the same samples was possible.

In binary iron–chromium alloys of high purity the presence of sigma can most easily be detected under the microscope by the typical cracking and fissuring which is associated with the sigma phase when present in a massive form. As is stated later, sigma in the ternary alloys is sometimes observed as a separate precipitated phase with certain definite etching characteristics, but it has not been found possible to distinguish with certainty between alpha and sigma when these phases are precipitated in the form of very small particles.

MICROSTRUCTURES OF THE ALLOYS

Details of the microstructures of the alloys after the various heat-treatments are given in Tables II, III, and IV. The following symbols have been used to denote the various phases, and have been chosen to conform with the notations used by Schafmeister and Ergang,[1] Bradley and Goldschmidt,[2] and Owen and Sully[3] :

- α The phase having the body-centred cubic lattice characteristic of iron.
- α_2 The supersaturated phase having the distorted body-centred lattice formed by the diffusionless transformation of the gamma phase in certain iron-rich alloys.
- α' The phase having the body-centred cubic structure characteristic of chromium.
- γ The phase having the face-centred cubic structure characteristic of nickel.
- σ The phase of complex structure observed in iron–chromium alloys.

There is no structural difference between the two stable alpha phases, and at high temperatures the equilibrium diagram contains only alpha and gamma phases, but it is convenient to distinguish these two alpha phase fields at lower temperatures where they are separated by other phase fields.

The presence of sigma in any one alloy could not always be distinguished by the use of any particular etching reagent, and the majority of the alloys were etched electrolytically either in an oxalic acid solution or in Vilella's reagent (aqua-regia in glycerine). Both these reagents usually attack sigma more readily than either the alpha or gamma solid solutions, and often the attack on sigma appears to be along certain preferred planes, only part of the particle being etched. The etched areas of the sigma phase have the appearance of lustrous wide pits. This results in a fairly characteristic appearance of the sigma particles, typical examples being shown in Figs. 4 and 20. A number of alloys consisted of alpha and gamma after quenching from a high temperature ; precipitation of sigma in some of these alloys occurred first in the alpha phase, considerably longer times being necessary for the precipitation in the gamma phase.

The microstructures reproduced in Figs. 1 to 46 serve to illustrate the precipitation of sigma in alloys which, after quenching from a high temperature, consisted either wholly of alpha, wholly of gamma, or of a mixture of alpha and gamma.

Precipitation of Sigma in wholly Ferritic Alloys

Figure 1 shows the high proportion of sigma existing in 1*F*30 after 190 days at 650° C. The structure is not unlike that which occurs in binary iron–chromium alloys ; the sigma phase would not be easily identified if fissuring—more readily discernible at high magnifications—were not present. In Figs. 2 and 3, alloy 3*F*25 is shown to consist of alpha, gamma, and a small amount of deeply etched sigma. The structure of another alloy, 5*F*30, in the ternary field at 650° C., is given in Fig. 4, the proportion of sigma being considerably greater in this alloy than in 3*F*25. The absence of sigma in 3*F*25 after 56 days at 800° C. is indicated in Fig. 5, the microstructure consisting of gamma areas in a matrix of alpha. It will be noted that the long crack avoids passing through the gamma areas. Similar cracks are frequently observed in alloys in which sigma has been formed. The microstructure of 5*F*22 after 156 days at 650° C. (Fig. 6) and after 128 days at 800° C. (Fig. 7) serves to illustrate how microscopic examination can usefully supplement the results of the X-ray examination. In addition to the dark-etching sigma phase in the specimen annealed at 650° C., the microstructure consists of alpha and gamma in approximately equal proportions ; the X-ray diffraction pattern shows only supersaturated alpha (α_2) and sigma. The two phases alpha and gamma are present also after annealing at 800° C. (Fig. 7), X-ray evidence indicating only supersaturated alpha (α_2) with a doubtful trace of gamma. The significance of the presence of a distorted alpha structure in the X-ray diffraction pattern of certain alloys is discussed in a later section. The microstructure in Fig. 8 shows the confused character of the distorted alpha structure. In some alloys this structure is revealed only by etching with certain reagents, such as Vilella's aqua-regia solution. Figures 9 to 11 illustrate the precipitation and growth of sigma in an alloy originally in the ferritic condition (Fig. 9), but which was in the ($\gamma + \sigma$) field at 650° C. Annealing for 14 days at 650° C. results in a finely dispersed precipitate (Fig. 10), but after 90 days the microstructure consists of a matrix of sigma in which are dispersed small particles of the gamma phase (Fig. 11). After 14 days at 800° C., the precipitate is coarser than after the 650° C. treatment for the same time, and the presence of fine fissures in the matrix, which avoid the precipitated particles, indicate that the matrix consists of sigma (Fig. 12). A similar structure to that in Fig. 12 is found in 5*F*34 after annealing at 650° C. for 190 days, but after 56 days at 800° C. the sigma is found distributed in large areas, as shown in Fig. 13.

Precipitation of Sigma in Alloys Consisting of Alpha and Gamma after Quenching from a High Temperature

The formation of sigma in an alloy consisting initially of alpha and gamma is illustrated in Figs. 14 to 16, which show alloy 12*F*20 annealed at 650° C. After 14 days (Fig. 14), sigma is precipitated only in the alpha phase, giving rise to a structure similar

Table II

MICROSCOPIC AND X-RAY EXAMINATION OF ALLOYS ANNEALED AT 650° C.

Alloy	Microscopic Examination of Lump Samples			X-Ray and Magnetic Observations			
	Quenched from High Temp.	After Annealing	Time of Anneal, days	Constitution*	Magnetic Props.†	Time of Anneal, days	Parameter of γ Phase, kX. at 20°C.
8F18	$\gamma + \alpha$	γ (changed to martensitic α_2), σ, some untransformed α	190	$\alpha_2 + \gamma + \bar{\sigma}$	S	265	3·5793
8F20	$\gamma + \alpha$	γ (changed to martensitic α_2), σ, possibly α	190	$\gamma + \sigma$	VW	265	3·5795
9F20	$\gamma + \alpha$	γ (changed to martensitic α_2), σ, possibly α	30	$\gamma + \sigma + \alpha_2$	S	265	3·5795
11F22	$\gamma + \alpha$	Precipitate of σ in α and γ phases	28	$\gamma + \sigma + \bar{\alpha}_2$	W	28	3·5796
11F40	$\gamma + \alpha$	Small particle of γ in σ matrix	14	$\sigma + \gamma$	VW	140	
12F20	$\gamma + \alpha$	Precipitate of σ in α and γ phases	90	$\gamma + \sigma\bar{\alpha}$	VW	4	3·5791
12F19	γ	Some σ precipitated	395	$\gamma + \bar{\sigma}$	VW	155	3·5792
13F18	γ	Unchanged γ	168	γ	VW	176	3·5782
13F45	$\alpha + \gamma$	Large proportion of σ, γ, and α	14	$\alpha + \gamma + \sigma$	..	4	..
14F15	γ	Unchanged γ	28	γ	VW	176	3·5771
14F20	γ	Some σ precipitated	162	$\gamma + \bar{\sigma}$	VW	155	3·5793
14F28	$\gamma + \alpha$	Much σ in original α areas	14	$\gamma + \sigma$	VW	16	3·5782
16F11	γ	Unchanged γ	30	$\gamma + \bar{\alpha}_2$	M	28	3·5745 (1)
17F23	γ	Some σ precipitated	90	$\gamma + \sigma$	VW	176	3·5783
18F5	α_2	α_2 unchanged	28	$\gamma + \alpha_2$	S	4	3·5736
19F20	γ	Unchanged γ	168	$\gamma + \sigma$	VW	176	3·5783
19F30	$\gamma + \alpha$	Much σ in original α areas	14	$\gamma + \sigma$	VW	16	..
20F0	α_2	Confused irreg. crystal structure	28	α_2	S	2	.. (2)
20F25	γ	Not annealed in lump form	..	$\gamma + \sigma$	VW	72	..
21F17	γ	Unchanged γ	30	γ	VW	176	3·5768
23F14	γ	Unchanged γ	28	γ	VW	4	3·5757
23F25	γ	Precipitate of fine particles in γ grains	14	$\gamma + \sigma$	W	176	3·5779
25F7	γ	Not annealed in lump form	..	γ	M	8	3·5735
26F24	γ	Striated precipitate, small amount of σ	90	$\gamma + \sigma$	W	176	3·5779
		σ more definite and massive	264				
26F26	γ	Striated precipitate, small amount of σ	90	$\gamma + \sigma$	W	176	3·5777 (3)
26F31	$\gamma + \alpha$	Precipitate in γ grains, some σ in α areas	42				
		σ more definite	234	$\gamma + \alpha$	M	168	3·5777
27F37	$\gamma + \alpha$	Acicular precipitate in both phases	14	$\gamma + \alpha$	VW	4	3·5765
30F19	γ	Unchanged γ	162	γ	M	155	3·5768
30F23	γ	Not annealed in lump form	..	$\gamma + \alpha$	W	60	3·5766
31F0	γ	Unchanged γ	28	γ	S	2	3·5838
31F26	γ	Precipitate of α in some γ grains	42	$\gamma + \alpha$	W	168	3·5765
32F14	γ	Unchanged γ	155	γ	M	155	3·5750
33F33	γ	Precipitate of fine particles in γ grains	168	$\gamma + \alpha$	M	155	3·5753 (4)
37F7	γ	Not annealed in lump form	..	γ	S	8	3·5798
35F12	γ	Unchanged γ	14	γ	S	4	3·5757 (1)
35F22	γ	Precipitate of α in some γ grains	42	$\gamma + \bar{\alpha}$	M	168	3·5757
35F25	γ	Not annealed in lump form	..	$\gamma + \alpha$	W	60	3·5749
37F29	γ	,, ,, ,, ,,	..	$\gamma + \alpha$	W	60	3·5739
38F18	γ	Unchanged γ	155	γ	M	155	3·5743
41F16	γ	Unchanged γ	42	γ	S	168	3·5741
41F24	γ	Unchanged γ	168	$\gamma + \alpha$	W	172	3·5737
42F23	γ	Precipitate in few isolated grains	390	$\gamma + \bar{\alpha}$	W	155	3·5737
45F19	γ	Unchanged γ	155	γ	M	155	3·5723
48F22	γ	Unchanged γ	28	γ	W	4	3·5687
50F30	γ	Not annealed in lump form	..	$\gamma + \alpha$	VW	8	3·5682 (4)
60F10	γ	,, ,, ,, ,,	..	γ	S	8	3·5634
0F30	α	,, ,, ,, ,,	..	$\alpha + \sigma$	S	140	
1F20	α	Unchanged α	190	α	S	265	
1F25	α	Unchanged α	190	$\alpha + \sigma$	S	265	
1F30	α	Large amount of fissured σ formed	190	$\alpha + \sigma$	S	265	
2F20	α	Unchanged α	190	$\alpha + \bar{\gamma}$	S	265	
2F25	α	Unchanged α	190	$\alpha + \bar{\gamma} + \sigma$	S	265	
2F28	α	Traces of γ particles	190	$\alpha + \sigma + \gamma$	S	28	
		(No change after 42 days)					
3F18	α	Small particles of γ precipitated	156	$\alpha + \bar{\gamma}$	S	168	
3F20	α	Similar to 3F18, but more γ	156	$\alpha + \alpha_2$	S	168	
3F25	α	Small amounts of σ and γ precipitated	190	$\alpha + \sigma + \gamma$	S	265	
4F25	α	Not annealed in lump form	..	$\alpha + \sigma + \gamma$	S	80	
5F18	$\alpha + \gamma$	$\alpha_2 + \alpha$, proportion of α_2 higher than that of γ before annealing	190	$\alpha + \alpha_2$	S	265	
5F20	$\alpha + \gamma$	$\alpha_2 + \alpha$, very small amount of σ	156	$\alpha + \alpha_2 + \bar{\sigma}$	S	168	
5F22	α	$\alpha + \gamma$, small amount of σ	156	$\alpha + \alpha_2 + \sigma$	S	168	
5F30	α	Not annealed in lump form	..	$\gamma + \sigma$	VW	80	
5F34	α	Small particles of γ in fissured σ matrix	190	$\gamma + \sigma$	VW	265	
5F38	α	Small amount of γ in a σ matrix	30	$\sigma + \gamma$	VW	28	
6F30	α	Not annealed in lump form	..	$\gamma + \sigma$	VW	80	
6F35	α	Few γ particles in σ matrix	90	$\sigma + \gamma$	VW	16	
9F37	$\alpha + \gamma$	Many fine γ particles in σ matrix	14	$\sigma + \gamma$	VW	140	

(1) Erratic parameter values.
(2) Complete breakdown of γ phase.
(3) Equilibrium poor after 650° C. anneal.
(4) Equilibrium poor

* $\bar{\alpha}$ = trace of α, $\bar{\gamma}$ = trace of γ, $\bar{\sigma}$ = trace of σ.
† Response to magnet: S = strong, M = medium, W = weak, V W = very weak.

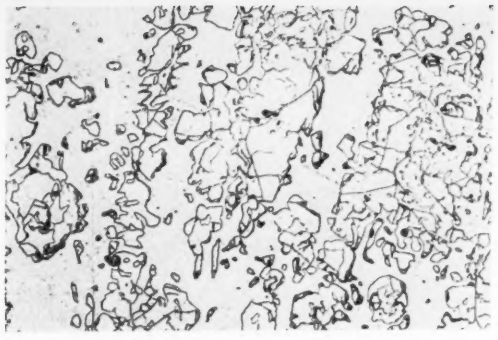


Fig. 1—1F30, 190 days at 650° C. × 150

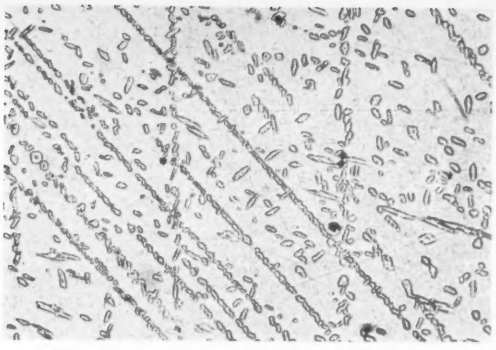


Fig. 2—3F25, 190 days at 650° C. × 150

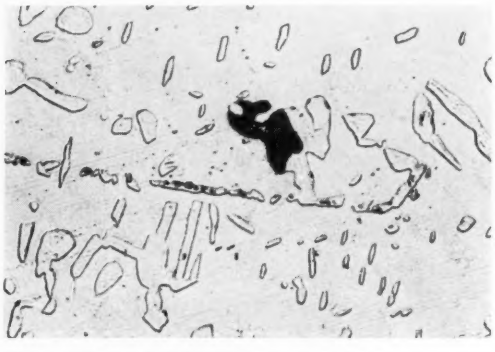


Fig. 3—3F25, 190 days at 650° C. × 500

Fig. 4—5F30, 16 days at 650° C. × 500

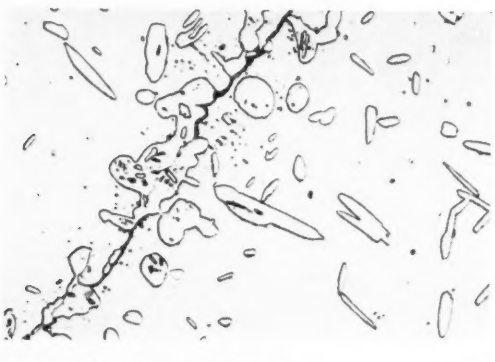


Fig. 5—3F25, 56 days at 800° C. × 500

Fig. 6—5F22, 156 days at 650° C. × 500

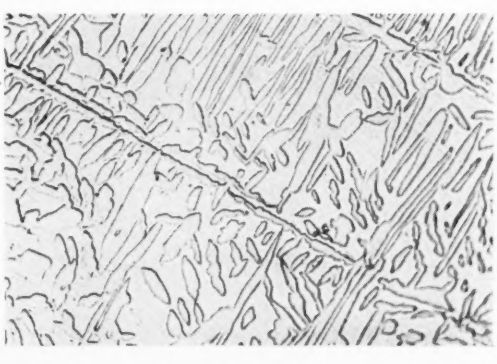


Fig. 7—5F22, 128 days at 800° C. × 500

Fig. 8—5F18, 190 days at 650° C. × 150

[*Rees, Burns and Cook*
[*To face p.* 328

Fig. 9—6F35, quenched at 1000° C. × 500

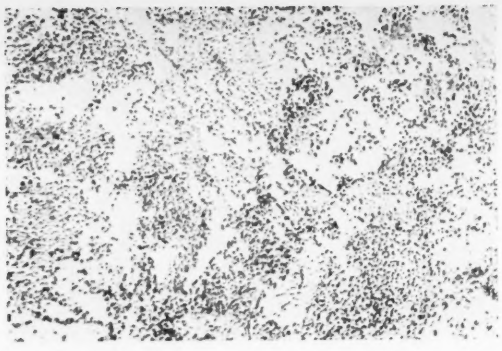


Fig. 10—6F35, 14 days at 650° C. × 500

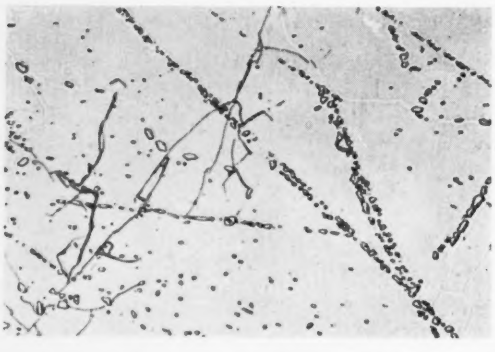


Fig. 11—6F35, 90 days at 650° C. × 500

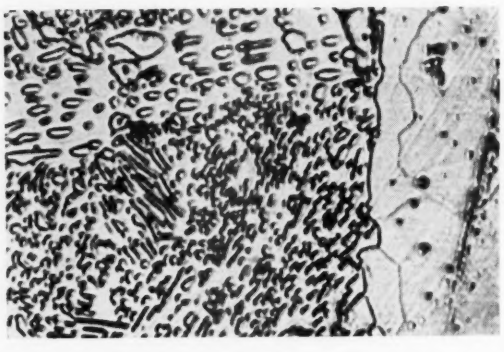


Fig. 12—6F35, 14 days at 800° C. × 1000

Fig. 13—5F34, 56 days at 800° C. × 500

Fig. 14—12F20, 14 days at 650° C. × 500

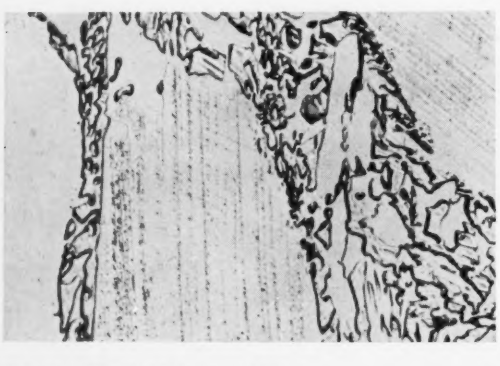


Fig. 15—12F20, 90 days at 650° C. × 500

Fig. 16—12F20, 90 days at 650° C. × 750

Rees, Burns and Cook]

Fig. 17—14F28, 14 days at 800° C. × 750

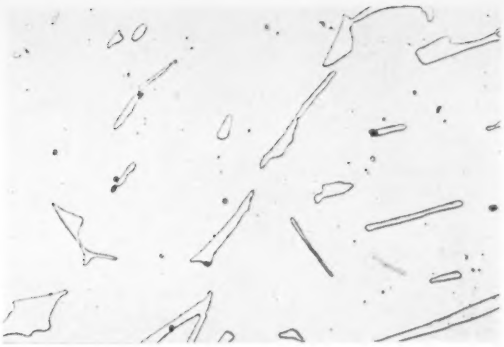


Fig. 18—8F18, quenched at 1000° C. × 150

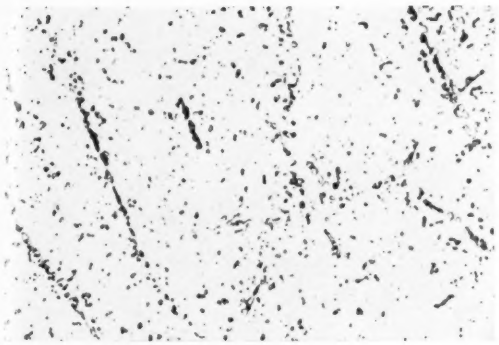


Fig. 19—8F18, 190 days at 650° C. × 150

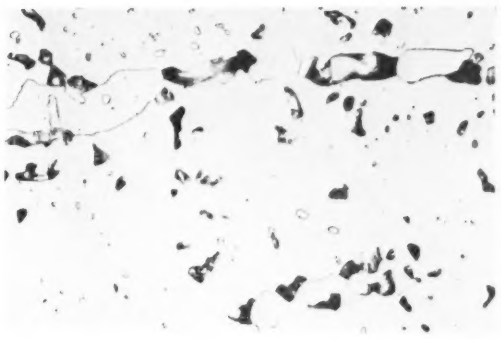


Fig. 20—8F18, 190 days at 650° C. × 500

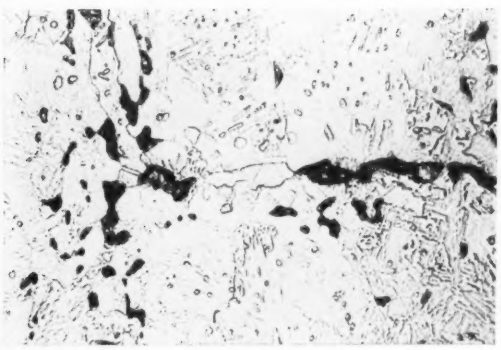


Fig. 21—8F18, 190 days at 650° C. × 500

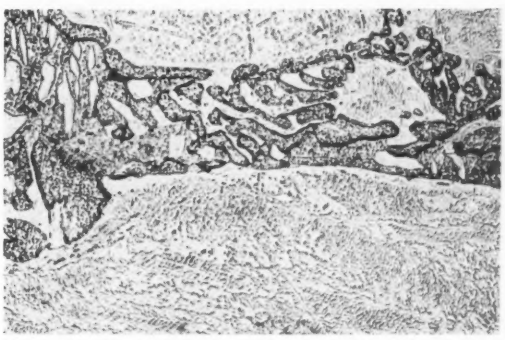


Fig. 22—8F20, 190 days at 650° C. × 500

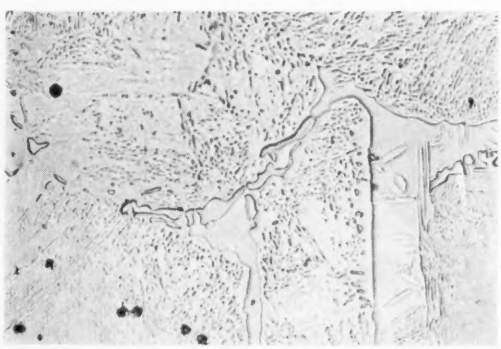


Fig. 23—9F20, 30 days at 650° C. × 500

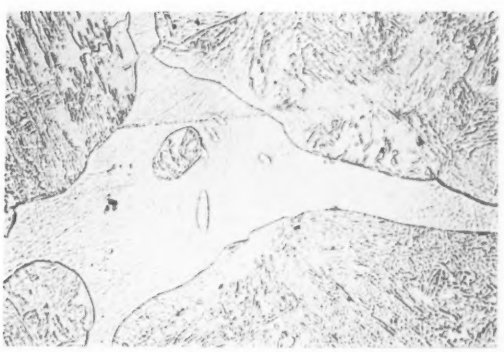


Fig. 24—8F18, 56 days at 800° C. × 500

[*Rees, Burns and Cook*

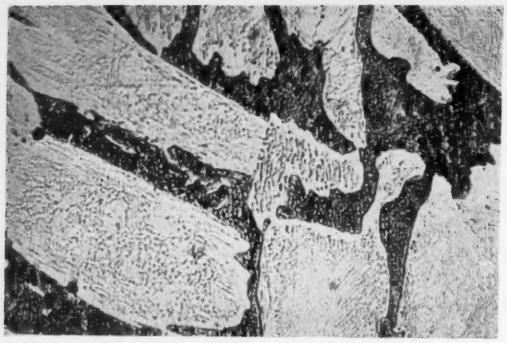


Fig. 25—8F20, 56 days at 800° C. × 500

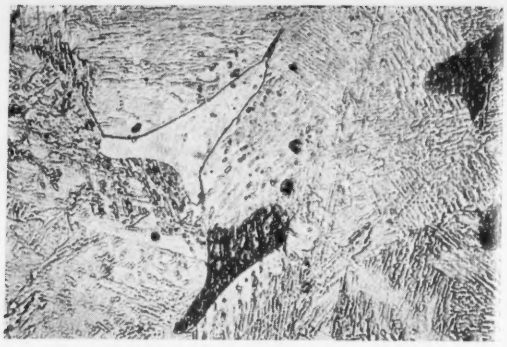


Fig. 26—9F20, 21 days at 800° C. × 500

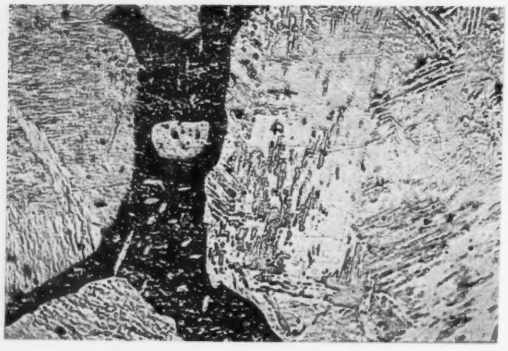


Fig. 27—9F20, 21 days at 800° C. × 500

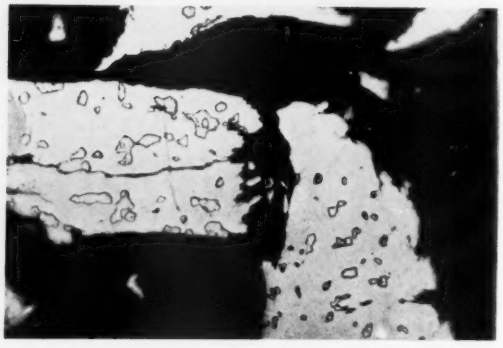


Fig. 28—8F18, 265 days at 650° C. × 500

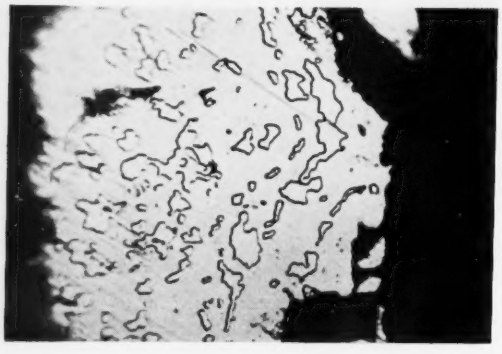


Fig. 29—8F20, 265 days at 650° C. × 1500

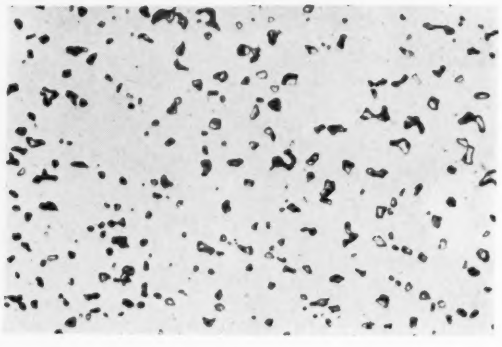


Fig. 30—8F18, 70 days at 650° C. × 500

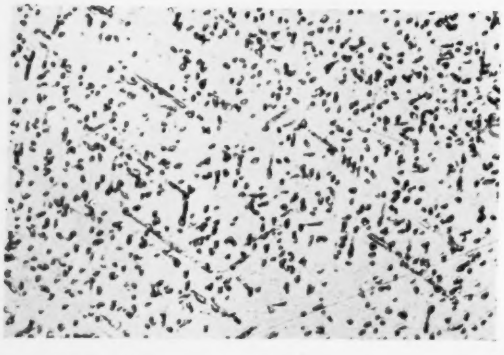


Fig. 31—26F26, 21 days at 800° C. × 500

Fig. 32—26F26, 86 days at 800° C. × 500

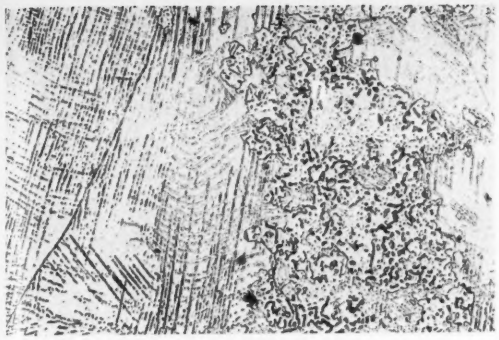


Fig. 33—26F26, 90 days at 650° C. × 150

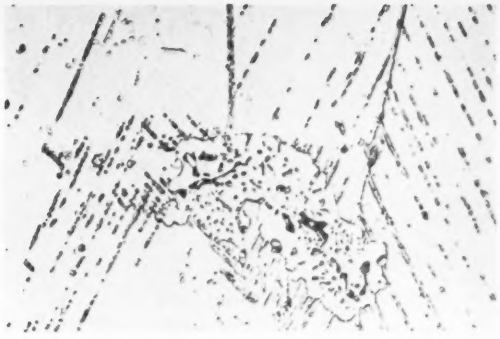


Fig. 34—26F26, 90 days at 650° C. × 500

Fig. 35—17F23, 30 days at 650° C. × 500

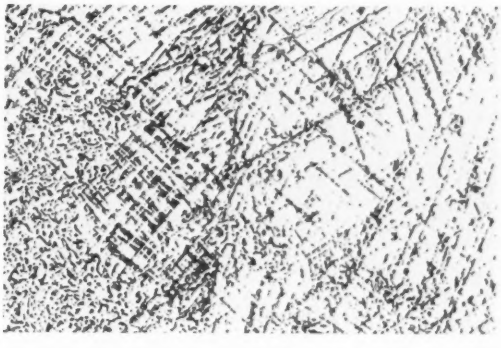


Fig. 36—17F23, 90 days at 650° C. × 150

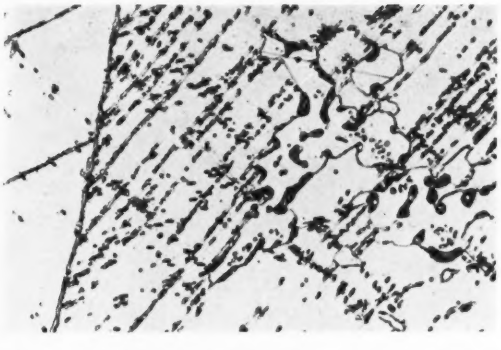


Fig. 37—17F23, 90 days at 650° C. × 500

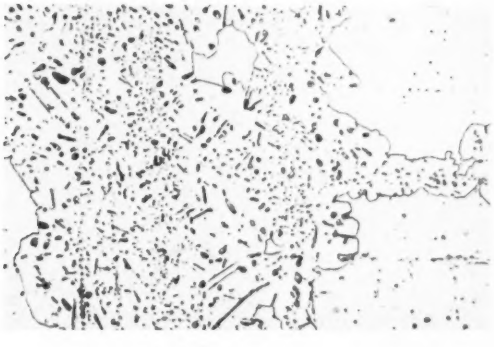


Fig. 38—31F26, 42 days at 650° C. × 500

Fig. 39—35F22, 42 days at 650° C. × 500

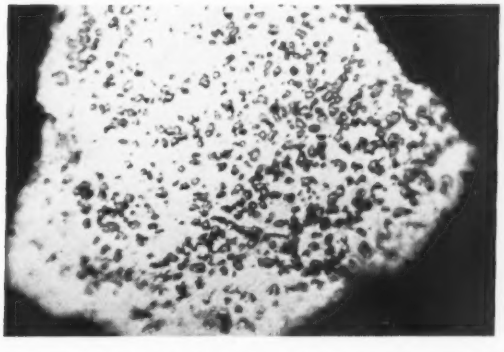


Fig. 40—17F23, 30 days at 650° C. X-ray powder sample × 1500

[*Rees, Burns and Cook*

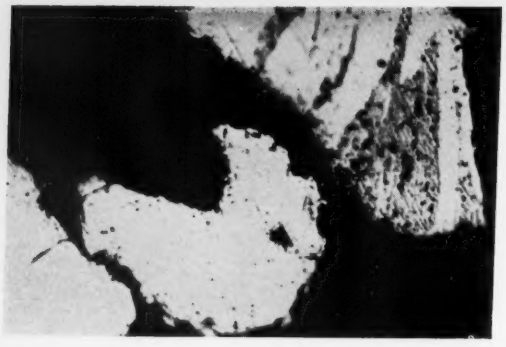


Fig. 41—14F28, quenched at 1000° C. ; X-ray powder sample from duplex alloy × 1500

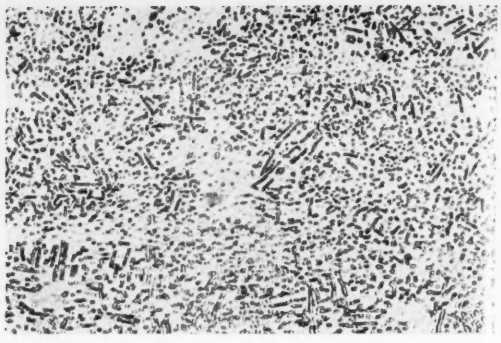


Fig. 42—26F26, 2 hr. at 800° C. × 500

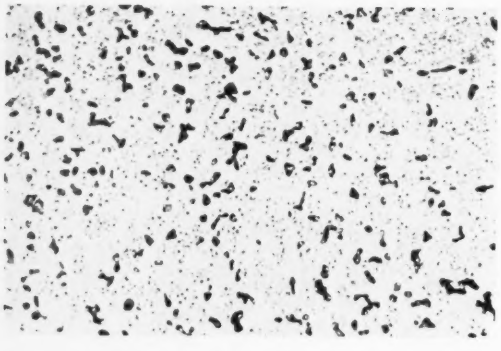


Fig. 43—26F26, 24 hr. at 800° C. × 500

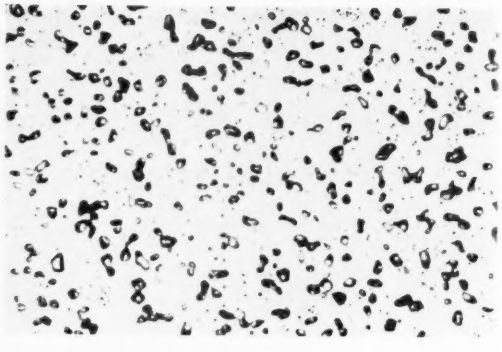


Fig. 44—26F26, 168 hr. at 800° C. × 500

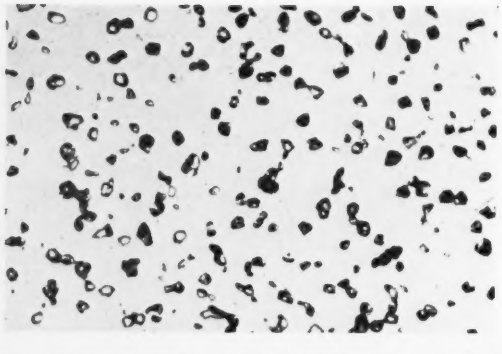


Fig. 45—26F26, 888 hr. (37 days) at 800° C. × 500

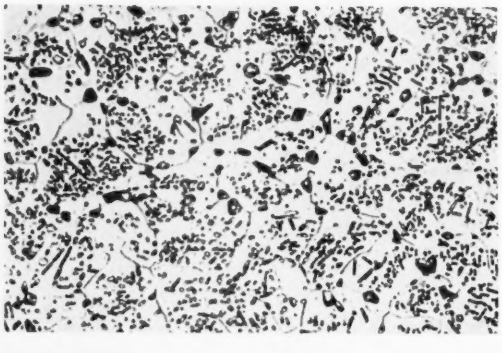


Fig. 46—31F26, 63 days at 800° C. × 500

Table III

MICROSCOPIC AND X-RAY EXAMINATION OF ALLOYS ANNEALED AT 800° C.

Alloy	Microscopic Examination of Lump Samples			X-ray and Magnetic Observations			
	Quenched from High Temp.	After Annealing	Time of Anneal, days	Constitution*	Magnetic Props.†	Time of Anneal, days	Parameter of γ Phase, kX. at 20° C.
8F18	$\gamma + \alpha$	Martensitic $\alpha(\alpha_2)$ and α. No σ	56	$\alpha_2 + \gamma$	S	86	3·5792
8F20	$\gamma + \alpha$	Martensitic α. Original α now duplex. No σ	56	$\gamma + \alpha + \alpha_2$	S	86	3·5797 (1)
9F20	$\gamma + \alpha$	,, ,, ,, ,,	21	$\gamma + \alpha + \alpha_2$	S	56	3·5802
11F22	$\gamma + \alpha$	Precipitation in γ and α phases, fissuring indicating σ	20	$\gamma + \bar{\sigma}\bar{\alpha}$	W	21	3·5812
11F40	$\alpha + \gamma$	Fissuring of matrix indicating σ formation	14	$\sigma + \gamma$	VW	56	..
12F20	$\gamma + \alpha$	No precipitate in γ phase, small islands of a constituent in α phase	14	$\gamma + \sigma$	VW	4	3·5811
12F19	γ	Unchanged γ	86	γ	VW	221	3·5795
13F18	γ	Unchanged γ	30	γ	VW	221	3·5782
13F45	$\alpha + \gamma$	Small particles of γ in σ matrix, possibly trace of α	14	$\gamma + \sigma$	..	4	..
14F15	γ	Unchanged γ	14	γ	VW	221	3·5771
14F20	γ	Unchanged γ	86	γ	VW	126	3·5794
14F28	$\gamma + \alpha$	Fissures in α phase and σ precipitated in γ phase	14	$\gamma + \sigma$	VW	4	3·5808
16F11	γ	Unchanged γ	30	$\gamma + \bar{\alpha}_2$	M	21	3·5745
17F23	γ	Fine precipitate in γ grains	30	$\gamma + \sigma$	VW	21	3·5809
18F5	α_2	Martensitic α (α_2)	14	$\gamma + \alpha_2$	S	4	3·5736
19F20	γ	Unchanged γ	14	γ	VW	221	3·5795
19F30	$\gamma + \alpha$	Original α areas transformed to σ	14	$\gamma + \sigma$	VW	4	
20F0	α_2	α_2 unchanged	14				(2)
20F25	γ	Not annealed in lump form	..	$\gamma + \sigma$	VW	126	3·5802
21F17	γ	Unchanged γ	14	γ	VW	221	3·5768
23F14	γ	Unchanged γ	14	γ	VW	4	3·5757
23F25	γ	Dense precipitate in γ grains	14	$\gamma + \sigma$	VW	221	3·5796
25F7	γ	Not annealed in lump form	..	γ	M	8	3·5735
26F24	γ	Unchanged γ after 21 days, fine precipitate, some coarse particles	124	$\gamma + (\bar{\sigma}\,?)$	VW	221	3·5794
26F26	γ	Coarse sigma particles, some α precipitate	124	$\gamma + \sigma$	VW	221	3·5795
26F31	$\gamma + \alpha$	Striated precipitate in γ matrix, probably some σ in α areas	124	$\gamma + \sigma$	VW	134	3·5791
27F37	$\gamma + \alpha$	Precipitate in both phases	14	$\gamma + \alpha$	VW	4	3·5793
30F19	γ	Unchanged γ	86	γ	W	126	3·5768
30F23	γ	Not annealed in lump form	..	γ	VW	60	3·5785
31F0	γ	Unchanged γ	14				..
31F26	γ	General fine precipitation, probably α	30	$\gamma + \bar{\sigma}$	VW	134	3·5795
32F14	γ	Unchanged γ	86	γ	M	126	3·5750
33F33	γ	Coarse precipitation, probably α	168	$\gamma + \alpha$	VW	221	3·5776
37F7	γ	Not annealed in lump form	..	γ	S	8	3·5798
35F12	γ	Unchanged γ	14	γ	S	4	3·5757 (1)
35F22	γ	Unchanged γ	30	γ	VW	134	3·5763
35F25	γ	Not annealed in lump form	..	$\gamma + \bar{\alpha}$	VW	60	3·5780
37F29	γ	,, ,, ,, ,,	..	$\gamma + \alpha$	VW	60	3·5772
38F18	γ	Unchanged γ	86	γ	M	126	3·5743
41F16	γ	Unchanged γ	30	γ	S	134	3·5741
41F24	γ	Unchanged γ	86	γ	VW	248	3·5742
42F23	γ	Unchanged γ	86	γ	VW	126	3·5741
45F19	γ	Unchanged γ	86	γ	M	126	3·5723
48F22	γ	Unchanged γ	4	γ	W	28	3·5696
50F30	γ	Not annealed in lump form	..	γ	VW	8	3·5730
60F10	γ	,, ,, ,, ,,	..	γ	S	8	3·5634
70F30	γ	,, ,, ,, ,,	..	γ	VW	3	3·5542
80F20	γ	,, ,, ,, ,,	..	γ	VW	3	3·5411
90F10	γ	,, ,, ,, ,,	..	γ	VW	3	3·5291
0F38	α	,, ,, ,, ,,	..	α	S	40	
1F20	α	Unchanged α	56	α	S	86	
1F25	α	Unchanged α	56	α	S	86	
1F30	α	Trace of fissured σ in α matrix	56	α	S	86	(3)
2F20	α	Unchanged α	56	α	S	86	
2F25	α	Unchanged α	56	$\alpha + \bar{\gamma}$	S	86	
2F28	α	Unchanged α	56	α	S	56	
3F18	α	Small quantity of γ precipitated	128	α_2	S	134	
3F20	α	,, ,, ,, ,,	128	α_2	S	134	
3F25	α	Coarse particles of γ precipitated	128	$\alpha + \bar{\gamma}$	S	86	
4F25	α	Not annealed in lump form	..	$\alpha + \gamma$	S	60	
5F18	$\alpha + \gamma$	Coarse particles of γ in α matrix	56	α_2	S	86	
5F20	$\alpha + \gamma$	,, ,, ,, ,,	128	$\alpha + \alpha_2$	S	134	
5F22	α	,, ,, ,, ,,	128	$\alpha + \alpha_2$	S	134	
5F30	α	Not annealed in lump form	..	$\alpha + \gamma(\bar{\alpha}\,?)$	S	60	
5F34	α	Massive areas of fissured σ, traces of α in γ matrix	56	$\gamma + \sigma(+\,\bar{\alpha}\,?)$	VW	56	
5F38	α	Particles mainly γ in σ matrix, possible trace of α	21	$\sigma + \gamma$	VW	56	
6F30	α	Not annealed in lump form		$\alpha + \gamma + \sigma$	S	60	
6F35	α	Precipitate of lamellae and particles in σ matrix	14	$\sigma + \gamma$	VW	56	
9F37	$\sigma + \gamma$	Fissured sigma formed in grain boundaries	14	$\sigma + \gamma$	VW	56	3·5826

(1) Erratic parameter values. (2) Complete breakdown of γ phase. (3) Faint additional lines at 800° C.

* $\bar{\alpha}$ = trace of α, $\bar{\gamma}$ = trace of γ, $\bar{\sigma}$ = trace of σ.
† Response to magnet: S = strong, M = medium, W = weak, VW = very weak.

to that of a eutectoid ; at this stage the alloy has not attained true equilibrium, and further changes occur with increased time of annealing. After 90 days, the sigma in the original alpha areas has become much coarser, and a finely divided precipitate—presumably sigma—appears in some of the gamma areas (Fig. 15). In a few isolated areas of the same specimen, development of sigma in both alpha and gamma areas had proceeded much further, resulting in the structure reproduced in Fig. 16. Heat-treatment for 14 days at 800° C. of alloy 14*F*28 transforms the alpha phase completely to sigma, a considerable proportion of sigma also being formed in the gamma areas ; the microstructure is seen in Fig. 17.

The three alloys 8*F*18, 8*F*20, and 9*F*20, which are close to the gamma triple point of the $(\alpha + \gamma + \sigma)$ ternary field at 650° C., could not be rendered wholly austenitic by quenching from a high temperature. The microstructures of these alloys, after annealing at 650° and 800° C., are of some interest, and illustrate the large effect that small differences in composition have on the response of the alloys to annealing. Figure 18 is representative of the microstructure of the three alloys as quenched from 1000° C. After annealing 8*F*18 for 190 days at 650° C., the original disposition of the alpha areas is not very evident but is indicated by the lines of sigma particles (Fig. 19). At higher magnification, parts of the original alpha areas can be seen to be unchanged (Figs. 20 and 21). In Fig. 21 the martensitic character of the original gamma areas has been revealed by etching in the aqua-regia/glycerine reagent. In alloy 8*F*20 the alpha areas have undergone a considerable change in structure, but their size and distribution have not changed very much (Fig. 22). The evidence for the presence of sigma lies in the dark-etching veins seen in the original alpha areas. The gamma areas have been etched to show that they have transformed to the distorted alpha phase on quenching from the annealing temperature. In alloy 9*F*20, the original alpha areas can still be discerned, but parts of them, as in 8*F*18, have been completely transformed to sigma, as shown in Fig. 23. Sigma does not appear to have been formed in the gamma areas of either 8*F*20 or 9*F*20. After 56 days at 800° C., no sigma was formed in 8*F*18, the alpha areas remaining apparently unchanged, Fig. 24. In 8*F*20, however, some change had occurred ; the alpha areas were readily etched, resulting in a structure (Fig. 25) which could not be resolved with an oil-immersion lens. A

Table IV

ADDITIONAL MICROSCOPIC AND X-RAY DATA FROM ALLOYS COLD WORKED BEFORE ANNEALING AT 650° C. AND 800° C.

Alloy	Quenched before Cold Work	Microscopic Features after 650° C. Anneal	Time of Anneal, days	X-ray Examination on Micro Samples	Microscopic Features after 800° C. Anneal	Time of Anneal, days	X-Ray Examination on Micro Samples Using Glancing-Angle Technique
12*F*19	γ	Uniform distribution of σ particles	96		Unchanged γ	63	
13*F*18	γ	Traces of σ particles in γ matrix	96		..		
17*F*23	γ	Mixture of fine precipitate and coarser σ particles	4	$\gamma + \sigma$	Same as at 650° C. but coarser	63	$\gamma + \sigma$
19*F*20	γ	Uniform distribution of σ particles	96		Unchanged γ	63	
21*F*17	γ	Unchanged γ	96		..		
26*F*24	γ	Fine precipitate and coarser σ particles in γ matrix	96		Few σ particles developed	63	
26*F*26	γ	,, ,, ,, ,,	96	$\gamma + \alpha + \sigma$	Only coarse σ particles in γ matrix	37	$\gamma + \sigma$
26*F*31	$\alpha + \gamma$	Coarse precipitated particles in both phases	113		..		
27*F*37	$\alpha + \gamma$	Coarse precipitate in both phases	46		Much coarser than at 650° C. and some particle darkening	52	$\gamma + \alpha$, no σ
30*F*23	γ	Well-defined areas containing striated precipitate	16		Unchanged γ	123	
31*F*26	γ	Uniform fine precipitation	60	$\gamma + \alpha$, no σ	Uniform precipitation, many elongated particles	63	$\gamma + \alpha +$ trace σ
33*F*33	γ	Uniform precipitation of coarse particles	16		..		
35*F*22	γ	Slight precipitation of dark-etching particles	60		Unchanged γ	63	
35*F*25	γ	Well-defined areas of striated precipitate	16	$\gamma +$ much α	Striated precipitate, large particles (σ ?) in clear matrix	123	
37*F*29	γ	General precipitation ; some dark-etching areas	6		Striated precipitate, large dark-etching particles (possibly σ)	123	
0*F*32	α	Small amount of fissured σ in grain boundaries	54		..		
0*F*38	α	Massive fissured σ round grain boundaries	54		..		
1*F*30	α	(Not examined)			Very small amount of fissured σ in α matrix	56	
1*F*25	α	No change from α			..		
2*F*20	α	Small γ areas developed in α	60		Small γ areas in α matrix	21	
2*F*28	α	Traces of γ in α matrix, traces of σ	70		Traces of γ	56	
2*F*25	α	Traces of second constituent, probably γ	60		..		
3*F*25	α	Small amount of σ, much γ in α matrix	60		Elongated particles of γ in α matrix	21	
4*F*25	α	Much σ with γ and α	42		,, ,, ,, ,,	21	
5*F*30	α	Mainly $\alpha + \gamma$, some σ	42		Particles of γ in α matrix, some σ	21	
5*F*34	α	(Not examined)			Small amount of γ in σ matrix	56	$\sigma + \gamma$ only
6*F*30	α	Much γ in σ matrix	42	$\gamma + \sigma$, no α	Particles of γ in α matrix, some σ	21	

similar change occurred in many of the alpha areas of 9*F*20, but in this alloy small areas in the easily etched ferrite appeared to be similar to the gamma matrix ; the structure is shown in Figs. 26 and 27.

It is thought that the change in the character of the alpha areas after annealing at 800° C. is due to the precipitation of a finely dispersed gamma phase. The microstructures of the powder samples of 8*F*18 and 8*F*20, after annealing for 265 days at 650° C., are shown in Figs. 28 and 29 to consist of small discrete particles of a constituent in a gamma matrix, complete elimination of the original distribution of the gamma and alpha phases having occurred. A similar elimination of the original gamma and alpha structure was obtained by annealing at 650° C. a heavily cold-worked sample in the form of a machined chip (Fig. 30). The appearance of the particles under the microscope—especially in 8*F*18—indicates that they are sigma. The proportion of sigma in 8*F*18, however, appears to be very much less than in the lump sample annealed at the same temperature, as will be seen by comparing Figs. 28 and 20.

The objections to the use of powder samples obtained from originally duplex alloys are referred to in the section dealing with X-ray examination, but it is convenient to call attention here to Fig. 41 which shows how the proportion of the two phases of a duplex alloy can vary in the separate particles of a powder sample.

Precipitation of Sigma in Alloys wholly Austenitic after Quenching from a High Temperature

In those alloys lying in the region of the $\gamma/(\gamma + \sigma)$ boundary which were wholly austenitic before the annealing treatments, any precipitated phase might naturally be expected to be sigma. In many alloys, however, the phase first precipitated is alpha in the form of fine particles which cannot readily be distinguished, by the etchants used, from finely precipitated sigma. This has caused some difficulty in interpreting the microstructures of alloys containing more than approximately 25% of nickel, as the alpha phase may be very persistent and may not change to sigma in a reasonable annealing period.

After 21 days at 800° C., a finely dispersed precipitate is observed in alloy 26*F*26 (Fig. 31). X-ray examination of the lump sample shows that the precipitate is predominantly alpha, though small amounts of sigma might be present without being detectable by X-ray methods. A longer treatment—86 days at 800° C.—results in a change in character and coarsening of the precipitate, as shown in Fig. 32, the precipitated phase etching in the manner characteristic of sigma.

The form of the precipitated phase in this alloy after 90 days' treatment at 650° C. is shown in Figs. 33 and 34. It first occurs along crystallographic planes and at this stage probably consists of alpha. Later, coarser and dark-etching particles appear in restricted patches, and these probably contain sigma, though at this stage it is often impossible to detect the presence of sigma by X-ray examination of the lump sample. This type of precipitation occurs in an alloy of lower nickel content, as shown in Figs. 36 and 37 which illustrate the microstructure of 17*F*23 after 90 days at 650° C. The microstructure in Fig. 37 should be compared with that in Fig. 40, which is a photomicrograph of an X-ray powder sample of the same alloy annealed at 650° C. for 30 days. The comparison suggests that the lump sample has not reached a state of complete equilibrium. Similar structures appear in alloys of higher nickel content, such as 31*F*26 and 35*F*22, which are shown in Figs. 38 and 39 after 42 days at 650° C. Sometimes the first precipitate is intergranular (Fig. 35).

Much light was thrown on these structures by the examination of a strip of alloy 26*F*26 which was cold rolled and repeatedly annealed at 800° C., with additional cold rolling between every stage of the annealing. The strip was examined after cumulative annealing times of 1, 2, 8, 24, 72, 168, 408, and 888 hr. (37 days). At first a uniformly distributed fine precipitate was produced (Fig. 42) ; on X-ray examination this proved to be alpha. Subsequently a few coarse dark-etching rounded particles appeared (Fig. 43), and these increased in number at the expense of the fine precipitate (Fig. 44) until, after 37 days, only the coarse particles remained (Fig. 45). X-ray examination then showed the alloy to consist of gamma and sigma only. It therefore appears that the initial fine and commonly acicular precipitate is generally alpha and that the coarse rounded particles which replace it are sigma. A strip of the same alloy which had been fully softened at 1000° C. and annealed alongside the cold-rolled strip for the same periods, contained much less sigma than alpha after 37 days.

The results of a number of other experiments on the annealing of cold-rolled strips are given in Table IV. The alloy 26*F*26 appeared to go through the same sequence of changes at 650° C. as at 800° C., but more slowly, and after 96 days at 650° C. it consisted of a mixture of gamma, alpha, and sigma, not unlike the structure shown in Fig. 44. On the other hand, the alloy 17*F*23, lower in nickel content than 26*F*26, consisted only of gamma and sigma after 4 days at 650° C. Alloys higher in nickel content, such as 31*F*26, 35*F*25, and 37*F*29, still contained alpha and possibly a trace of sigma, after 123 days at 800° C. (Fig. 46), although these alloys have not yet shown any trace of sigma on annealing at 650° C. It is possible that some of these alloys with nickel contents in the neighbourhood of 26% are actually in the $(\gamma + \alpha' + \sigma)$ field at 800° C., though longer annealing treatments will be necessary to confirm this point. The probable explanation of the failure to observe sigma at 650° C. is that the changes at 650° C. are similar to, but very much slower than, those at 800° C.

Hardening Effect of Sigma on Austenitic Alloys

Experiments were made to determine whether the presence of finely precipitated sigma had a marked effect on the hardness of alloys whose compositions were in the region of the $\gamma/(\gamma + \sigma)$ boundary, and which were known to develop small quantities of sigma. Small cold-worked samples of alloys 17*F*23 and 26*F*26 were annealed for increasing times and were examined between successive annealings for changes in hardness and microstructure. The results are

Table V

HARDNESS CHANGES DURING THE PRECIPITATION OF SIGMA IN AUSTENITIC ALLOYS

Alloy	Treatment	Total Time, hr.	Hardness, D.P.H.	Microstructural Features
17*F*23	Cold rolled	...	327	
	Cold rolled, annealed at 650° C.	5 min.	284	
		2	267	Precipitate of fine particles
		24	250	Some coarser particles
		96	212	
		216	185	Structure confirmed as $\gamma + \sigma$ by X-ray
	Cold rolled, softened by anneal at 1000° C.	...	98	Wholly austenitic
14*F*15	Cold rolled	...	347	Values indicate degree of softening obtained with austenitic alloy which does not undergo constitutional changes at 650° C.
	Cold rolled, annealed at 650° C.	216	116	
	Cold rolled, softened by anneal at 1000° C.	...	89	
26*F*26	Cold rolled	...	328	
	Cold rolled between anneals at 800° C.	1	187	
		2	...	Precipitation as in Fig. 42 ($\gamma + \alpha$)
		24	185	Coarse particles forming as in Fig. 43
		72	177	,, ,, ,, ,, ,,
		168	178	,, ,, ,, ,, ,,
		408	180	Mainly coarse particles in clear matrix
		888	172	Structure confirmed as $\gamma + \sigma$ by X-ray (Fig. 45)
	Cold rolled, softened by anneal at 1000° C.	...	105	Wholly austenitic
26*F*24	Cold rolled	...	265	
	Cold rolled between anneals at 800° C.	1	...	
		2	...	Wholly austenitic
		24	116	
		72	119	No change
		168	111	
		408	112	
		888	111	Few σ particles visible
	Cold rolled, softened by anneal at 1000° C.	...	99	Wholly austenitic

shown in Table V, from which it will be seen that in each case there is at first a drop in hardness due to the removal of cold work, but with increased annealing times and consequent replacement of the finely precipitated alpha by precipitated sigma, the hardness value does not rise again.

Alloy 26*F*26 (cold-worked), after annealing for 37 days at 800° C., retained a Vickers pyramid hardness value of 172, compared with 105 when fully softened, although its microstructure (Fig. 45) shows only the relatively coarse sigma particles in a matrix of gamma.

In contrast to these values obtained from alloys near the $\gamma/(\gamma + \sigma)$ boundary, alloy 13*F*45, which had a hardness value of 362 (D.P.H.) in the softened condition, after annealing at 800° C. and reaching equilibrium as $(\sigma + \gamma)$, attained a hardness of 883 (D.P.H.).

X-RAY EXAMINATION

Tables II and III summarize the results of the X-ray examination of powder samples. The annealing times, phases observed after annealing, the lattice parameter of the gamma phase, and the results of a rough test of the magnetic properties of the annealed powder, are given.

This method of examination has the advantage that the constitutional changes in the cold-worked powders are appreciably faster than in the annealed lump samples; moreover, the lattice-parameter value and the appearance of the diffraction lines give an idea of the state of equilibrium in the powder. So far as was practicable, annealings were continued until the sharpness of the lines indicated that a state of equilibrium had probably been reached. It is important to bring the alloys to a single-phase condition, if possible, before preparing the powder. If the alloys have a coarse duplex structure at this stage, different pieces of powder may contain widely differing proportions of the phases, as is illustrated in Fig. 41, and hence may have widely differing compositions. Misleading results may be obtained from such powders. Consequently greater reliance is to be placed on those data in Tables II and III which refer to initially homogeneous alloys.

Iron–nickel alloys with up to 26% of nickel, when quenched from the gamma state, undergo a partial or complete breakdown by diffusionless transformation to the distorted body-centred cubic structure (α_2). In the ternary system the addition of chromium was found to reduce the range of nickel composition over

which this occurred, and alloys 14*F*15, 12*F*20, and 5*F*30 were free from this effect. Nevertheless, its occurrence complicated the definition of the $(\alpha + \gamma)/\gamma$ and $(\alpha + \gamma + \sigma)/(\gamma + \sigma)$ boundaries, although the detection of the presence of sigma was unaffected. Frequently only a partial breakdown occurred, and it was then possible to distinguish the presence of the undistorted alpha phase (α) formed at the annealing temperature, except in alloys containing less than 4% of nickel, in which the distorted alpha (α_2) could not be distinguished from the undistorted alpha (α).

All the powder work was carried out with two Debye Scherrer cameras of 10 cm. dia. The parameter values were extrapolated from values calculated from the dimensions of the camera and film measurements alone, all films being corrected directly for shrinkage errors. Lattice-parameter measurements of pure silver and gold showed satisfactory agreement with the accepted values. Unfiltered chromium radiation was used. The parameter values were corrected for room-temperature variations by adopting a mean coefficient of expansion for alloys not subject to magnetic transition at or near room temperature, and a standard temperature of 20° C. was adopted to minimize the necessary corrections. The values are estimated to be accurate to $\pm$ 0·00015 kX. units.

Experiments on alloys 17*F*23 and 26*F*26, both of which lay within the $(\gamma + \sigma)$ field at 650° C., illustrated the great differences in the rates at which the alloys reached equilibrium. Two sets of powder from each alloy were annealed at 650° C., one in the cold-worked condition, and the other after annealing at 1000° C. so as to restore the strain-free austenitic condition. Detailed observations on these powders are given below :

Development of the Sigma Phase in Alloys 17*F*23 *and* 26*F*26 *at* 650° *C.*

Time at 650° C.	Phases Present*	Lattice Parameter of γ Phase at 20° C., kX. Units
Alloy 17*F*23, cold-worked powder		
5 min.	$\gamma + \sigma + \alpha$	3·5794
30 min.	$\gamma + \sigma$	3·5792
1 hr.	$\gamma + \sigma$	3·5793
24 hr.	$\gamma + \sigma$	3·5792
48 hr.	$\gamma + \sigma$	3·5791
30 days	$\gamma + \sigma$	3·5789
Alloy 17*F*23, strain-free powder		
Nil	γ	3·5824
30 hr.	γ	3·5820
5 days	γ	3·5822
10 days	γ	3·5824
17 days	$\gamma(+ \bar{\sigma}\ ?)$	3·5819
22 days	$\gamma + \bar{\sigma}$	3·5817
30 days	$\gamma + \bar{\sigma}$	3·5819
125 days	$\gamma + \bar{\sigma}$	3·5816
Alloy 26*F*26, cold-worked powder		
5 min.	$\gamma + \alpha$	...
1½ hr.	$\gamma + \alpha$	3·5783
2 days	$\gamma + \alpha$	3·5781
4 days	$\gamma + \alpha + \bar{\alpha}$	3·5782
8 days	$\gamma + \sigma + \alpha$	3·5778
30 days	$\gamma + \sigma(+ \bar{\alpha}\ ?)$	3·5778
Alloy 26*F*26, strain-free powder		
Nil	γ	3·5832
14 days	$\gamma + \bar{\alpha}$	3·5818
24 days	$\gamma + \alpha$	3·5819
30 days	$\gamma + \alpha$	3·5820
142 days	$\gamma + \alpha$	3·5813

* $\bar{\alpha}$ = trace of α, $\bar{\sigma}$ = trace of σ.

Alloy 17*F*23 in the strain-free condition developed the sigma structure in approximately 20 days without any evidence of the prior production of the alpha phase. Alloy 26*F*26, however, precipitated the alpha phase in under 14 days, and this alpha persisted unchanged after four months at 650° C. The alpha-phase lines remained diffuse, and it was clear that the alloy was well removed from its equilibrium condition. That this alloy does contain sigma as an equilibrium constituent at 650° C. has been demonstrated by the microscopic examination of lump material (*see* Fig. 34). It is possible that the sigma phase was present, but in too small a proportion to be detected by X-ray techniques.

With the cold-worked powder specimens, an appreciable amount of the sigma phase was developed in alloy 17*F*23 after only 5 min. annealing at 650° C., whilst sigma was first observed in the cold-worked powder specimens of 26*F*26 after 4 days, accompanied in each case by the alpha phase (α_2) formed on cold working. The effect of cold work has been to reduce the time necessary to precipitate the sigma phase by a factor greater than 6×10^3, and if a similar factor is operative for alloy 26*F*26 the sigma phase would not attain a significant proportion in the annealed alloy until after 70 years at 650° C. This aspect is of obvious practical importance. The observations are also of importance in deriving the equilibrium boundaries at compositions which are even more sluggish in reaction than alloy 26*F*26. In such alloys the metastable alpha phase, whether formed initially by cold work or by precipitation, may be very persistent, and there can be considerable doubt as to whether the alpha constituent of an alloy is stable or metastable.

The lattice parameters of the cold-worked powders after short annealings at 650° C. were in both cases appreciably lower than those of the corresponding strain-free powders, which suggested that a significant amount of precipitation occurred either during the cold working or in the very early stages of annealing.

The constitution diagrams for 650° C. and 800° C., as derived from the X-ray phase analysis of powders, are given in Figs. 47(*a*) and 48(*a*). The corresponding diagrams derived from microscopic analysis are given in Figs. 47(*b*) and 48(*b*). At 800° C. satisfactory equilibrium was attained in the powder samples for all critical alloys, as judged by the definition of the gamma diffraction lines and the stability of the parameter values. In alloys developing a high proportion of precipitate, particularly of sigma phase, the secondary lines were somewhat diffuse. At 650° C., while satisfactory equilibrium was obtained in critical alloys containing up to roughly 26% of nickel by annealing for times of the order of six months, beyond this point there was a marked increase of sluggishness, and alloys developed the metastable alpha structure. In view of the evidence of slow formation of sigma after the appearance of alpha, and the very slow reaction rates in these alloys, the observations cannot be regarded as providing any evidence of the position of the $(\gamma + \sigma + \alpha')$ and $(\gamma + \alpha')$ fields.

An estimate of the position of the $\gamma/(\gamma + \sigma)$ and $\gamma/(\gamma + \alpha')$ boundaries can also be made from the lattice-parameter measurements of the gamma phase.

It is necessary first to establish the isoparameter contours in the pure gamma field ; this has been done by plotting sets of parameter/composition curves both for constant nickel contents and for constant chromium contents, and interpolating to produce consistent families of curves. The observations of Owen and Sully[3] on iron–nickel alloys, and of Jette[4] and his co-workers, on nickel–chromium alloys, have been incorporated in these diagrams, and the resultant isoparameter contours are reproduced in Figs. 49 and 50. The observed lattice parameters after annealing at 650° C. are plotted in Fig. 49, and after annealing at 800° C. in Fig. 50. For each temperature those alloys in which the observed parameters fall appreciably below the value indicated by the contours are those in which precipitation of a second phase has occurred, and they lie on one side of the indicated gamma phase boundary lines. The phase boundaries so established agree with those given in Figs. 47 and 48. Detailed consideration of the figures suggests that there may be a slight break in the 800° C. gamma phase boundary around 30 to 35% of nickel, and a similar break in the 650° C. boundary at 40% of nickel. These breaks may indicate the position of the gamma corner of the $(\gamma + \sigma + \alpha')$ field.

The isoparameter contours in the pure gamma field exhibit a marked discontinuity where they intersect the magnetic transition boundary, which is indicated by a broad loop in Figs. 49 and 50. All alloys situated within this loop showed ' strong ' or ' medium ' response when tested with a strong permanent magnet, whereas alloys outside the loop (with the exception of those low-nickel alloys in which the gamma phase underwent diffusionless transformation on cooling) gave a ' very weak ' response when annealed at 800° C. When annealed at 650° C. some of the nickel-rich alloys outside the loop gave a somewhat stronger response, described as ' weak.' This effect has been observed in alloys, such as 26*F*26, which had developed a considerable proportion of sigma phase and which contained little or no metastable alpha, as well as alloys, such as 31*F*26, in which the secondary constituent was the metastable alpha. This would suggest that the increased magnetism is associated with the gamma phase and that the Curie-point boundary may actually intersect the phase boundary at 650° C. in these regions. The range of compositions in which the gamma phase was ferro-magnetic at room temperature was more extensive than the observations of Chevenard and other workers would suggest.

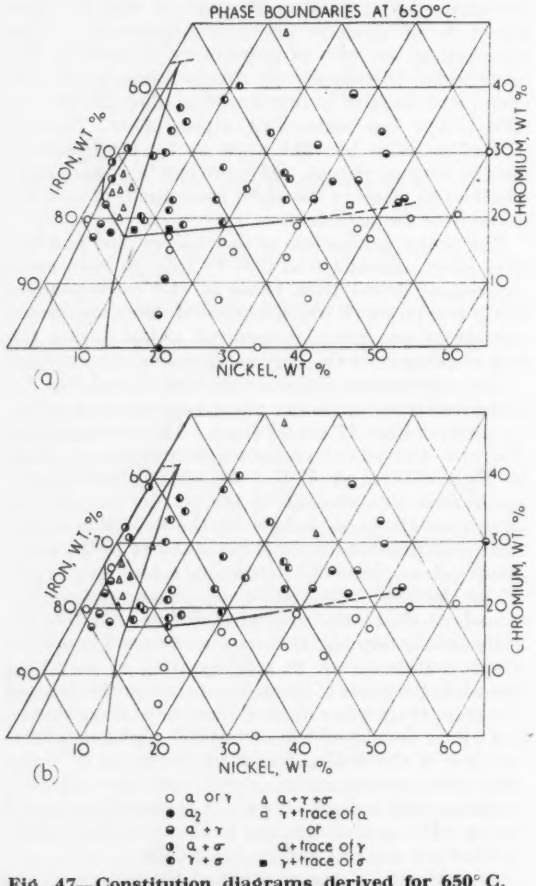


Fig. 47—Constitution diagrams derived for 650° C.
(*a*)—X-ray observations
(*b*)—Microscopic observations

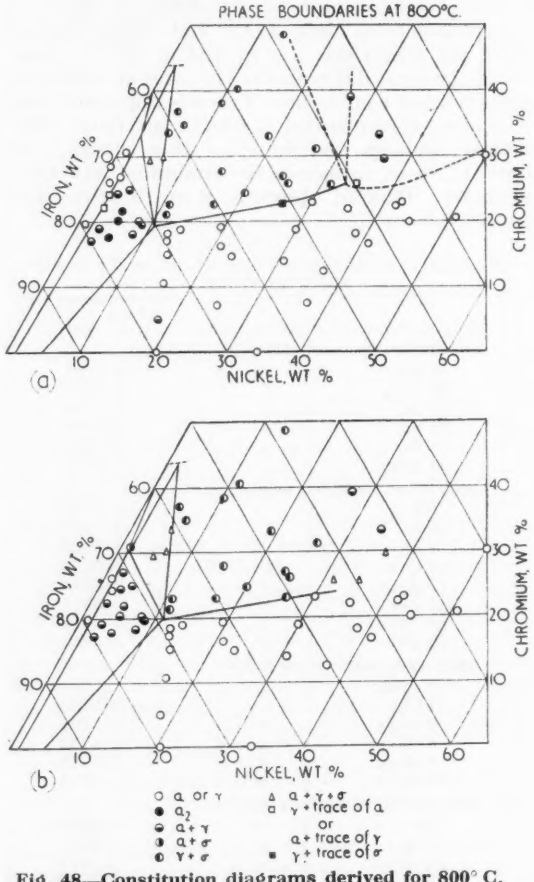


Fig. 48—Constitution diagrams derived for 800° C.
(*a*)—X-ray observations
(*b*)—Microscopic observations

DISCUSSION ON THE CONSTITUTION DIAGRAM AT 650° AND 800° C.

The phase fields in the binary iron–chromium system and the iron–nickel system, as given by Cook and Jones[5] and by Owen and Sully,[3] respectively, have been incorporated in Figs. 47 and 48.

Phase Field Boundaries at 650° C.

The agreement between the X-ray and the microscopic examination of the alloys annealed at 650° C. was generally very close, disagreement being confined to two or three alloys near to the phase field boundaries. Considering the three-phase field, the directions of the $(\alpha + \sigma)/(\alpha + \gamma + \sigma)$ and of the $(\alpha + \gamma)/(\alpha + \gamma + \sigma)$ boundaries have been fixed within narrow limits by the X-ray examination of alloys 1*F*25, 2*F*20, 2*F*25, 3*F*25, and 5*F*20. Consequently the alpha triple point is well defined. Alloy 1*F*25 contained less than 5% of sigma and no trace of gamma, while alloy 2*F*20 showed a trace of sigma and about 5% of gamma. The sigma and gamma contents of 2*F*25, 3*F*25, and 4*F*25 confirmed the direction of the $(\alpha + \gamma)/(\alpha + \gamma + \sigma)$ boundary fixed by alloy 5*F*20. These alloys were single-phase (α) at 1000° C., and were all subjected to very long annealing treatments. The positive X-ray evidence of the presence of sigma in alloys 1*F*25 and 2*F*25 has been given more weight when fixing the boundary than the negative evidence of its absence afforded by the microscopic examination of annealed lump samples. Some difficulty arose in fixing the gamma point of the three-phase triangle, since it lies in a region where the alloys cannot be rendered in a single-phase condition. Adequate fixation, however, was obtained from the direction of the $(\alpha + \gamma)/(\alpha + \gamma + \sigma)$ boundary and by extrapolation of the $\gamma/(\gamma + \sigma)$ boundary. This extrapolation has been determined principally by alloy 8*F*18 which, although duplex in the initial lump, developed only a very uncertain trace of sigma in the powder sample after 265 days' annealing, and certainly far less than was observed in 5*F*20. The presence of sigma was confirmed by microscopic examination. Alloys 8*F*20 and 9*F*20 showed no trace of the undistorted alpha structure and contained 5% or less of the sigma phase. The position of the sigma triple point and the direction of the $(\gamma + \sigma)/(\alpha + \gamma + \sigma)$ boundary are tentative, but cannot be much in error unless the range of chromium content, over which the sigma field is formed, is markedly altered by quite small additions of nickel. The high proportion of gamma in alloy 5*F*34, after 265 days at 650° C., suggests that this may be the case.

The $\alpha/(\alpha + \gamma)$ and $\gamma/(\alpha + \gamma)$ boundaries have been fixed by the alpha and gamma triple points and by the boundary limits in the iron–nickel system determined by Owen and Sully. Alloys 3*F*18, 3*F*20, and 5*F*18 contain the supersaturated alpha structure, but since the distortion in alloys with nickel contents of 5% or less is not marked, it is not possible, from the X-ray examination, to estimate the probable proportions of alpha and gamma which existed at the annealing temperature. The microscopic observations are in agreement with the boundaries which have been drawn.

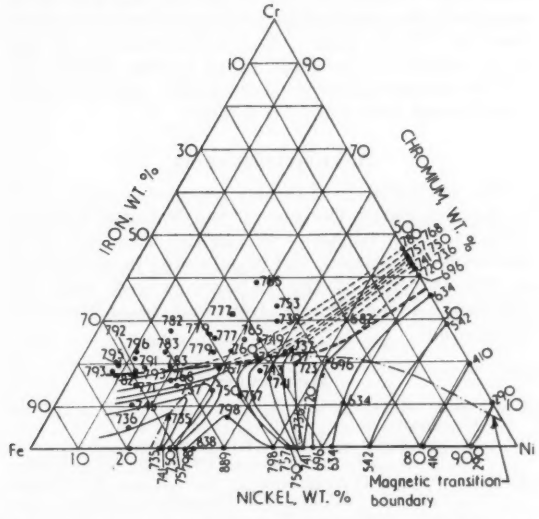


Fig. 49—Lattice-parameter values of gamma phase in alloys annealed at 650° C.

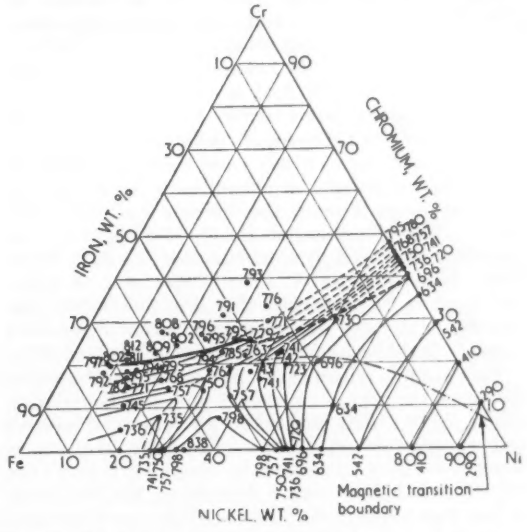


Fig. 50—Lattice-parameter values of gamma phase in alloys annealed at 800°C.

The $\gamma/(\gamma + \sigma)$ boundary is considered to be fixed with an accuracy of $\pm 0{\cdot}5\%$ of chromium up to approximately 25% of nickel, but beyond this nickel content the exact equilibrium structures are in some doubt, and it is not possible to define the position of the gamma triple point of the $(\alpha' + \gamma + \sigma)$ ternary field. The slight break in the gamma phase field boundary at about 40% of nickel suggested by the lattice-parameter values, possibly indicates the position of the gamma triple point.

Phase Field Boundaries at 800° C.

Although in the X-ray film the definition of the lines of alloys annealed at 800° C. indicated that

satisfactory equilibrium had been attained in the annealing times employed, the fixing of the phase field boundaries is somewhat less definite than at 650° C., and there was a greater divergence in the results of the X-ray and the microscopic examination of some of the alloys than was met with at 650° C. Considering in the first place the $(\alpha + \gamma + \sigma)$ ternary field, the position of the alpha triple point is fixed at 2% of nickel and 33% of chromium from the X-ray results with powder samples, and at 1½% of nickel and 30% of chromium from the results of the microscopic examination of lump samples. This difference is accounted for by a very definite trace of sigma in the microstructure of 1*F*30, whilst no sigma was detected in the X-ray powder specimen. There were, however, a number of additional unidentified lines in the X-ray spectrum of the alloy. There was also a second constituent seen under the microscope in 2*F*28, but the X-ray examination showed only alpha and a few additional unidentified lines. Some evidence was obtained, during the examination of some of the alloys, that the ternary field was very rapidly becoming narrow in the region of 800° C., and one alloy in particular, 5*F*34, developed different amounts of alpha—in addition to gamma and sigma—in different specimens, although no significant variations in chemical composition of the different specimens was found. The direction of the $(\gamma + \sigma)/(\alpha + \gamma + \sigma)$ boundary was fixed by the behaviour of alloys 5*F*34, 5*F*38, 6*F*30, and 6*F*35. The proportions of alpha and sigma in 6*F*30 indicate that it is definitely in the ternary field, and the varying behaviour of 5*F*34 suggests that this alloy is very close to the boundary.

The gamma triple point was fixed by the intersection of the $\gamma/(\gamma + \alpha)$ and the $\gamma/(\gamma + \sigma)$ boundaries. Alloys 8*F*18, 8*F*20, and 9*F*20 all contained gamma and supersaturated alpha. Normally the supersaturated alpha structures in alloys with less than 9% of nickel show severe distortion ; the resolution of the high-angle alpha lines in 8*F*20 and 9*F*20 indicates that appreciable amounts of alpha were present at the annealing temperature. Since these alloys were duplex when quenched from 1000° C. or higher, it might be argued that the alpha present at 800° C. was derived from that at 1000° C. Microscopic examination of powder samples of these alloys annealed at 650° C. gave no indication of serious composition differences, and microscopic examination of lump specimens of these alloys annealed at 800° C. for 56 days showed no tendency to approach a single-phase condition. It is therefore fairly certain, from both X-ray and microscopic examination, that 8*F*18, 8*F*20, and 9*F*20 are within the $(\alpha + \gamma)$ field at 800° C.

The extrapolation of the $\gamma/(\gamma + \sigma)$ boundary was guided by observations on alloys 11*F*22 and 12*F*20. Both these alloys were grossly duplex as quenched from 1000° C. Alloy 11*F*22, however, showed little or no sigma, and alloy 12*F*20 showed only a small amount. The small trace of alpha observed in the X-ray film of alloy 11*F*22 may have been due to a tendency for the gamma phase to transform on quenching, or more probably to the incomplete transformation to sigma. These two alloys are located at a relatively wide part of the $(\alpha + \gamma)$ field at 1000° C., compared with alloys 8*F*18 and 8*F*20, and consequently the composition variations in powder particles may be correspondingly greater. The appearance of only very small amounts of sigma in these alloys, together with the trace of partially formed sigma in alloy 17*F*23, justifies the location of the boundary.

From the gamma triple point of the $(\alpha + \gamma + \sigma)$ field, the $\gamma/(\gamma + \sigma)$ boundary extends to at least 31% of nickel, and is appreciably displaced away from the iron–nickel base line, compared with its position at 650° C. The data available are insufficient to enable the position of the gamma triple point of the $(\alpha' + \gamma + \sigma)$ ternary field to be given with any degree of certainty, but the break in the gamma boundary at about 33% of nickel indicated by parameter values (Fig. 50) suggests that it may be near to 33% of nickel and 25% of chromium. In Fig. 48(*a*) the boundaries of the $(\alpha' + \gamma + \sigma)$ ternary field have been tentatively drawn to agree with the evidence obtained from the X-ray work. The microstructure of the lump samples of the three alloys 31*F*26, 35*F*25, and 37*F*29, cannot, however, be reconciled with the X-ray examination of the powder samples.

Acknowledgments

The work described in this paper was carried out in the Metallurgy Division of the National Physical Laboratory on behalf of the British Iron and Steel Research Association, and is published by permission of the Director of the Laboratory and with the approval of the British Iron and Steel Research Association.

References

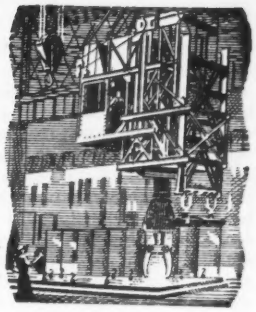
1. P. SCHAFMEISTER and R. ERGANG : *Archiv für das Eisenhüttenwesen*, 1939, vol. 12, p. 459.
2. A. J. BRADLEY and H. J. GOLDSCHMIDT : *Journal of The Iron and Steel Institute*, 1941, No. II, p. 273P.
3. E. A. OWEN and A. H. SULLY: *The London, Edinburgh, and Dublin Philosophical Magazine and Journal of Science*, 1939, vol. 27, p. 614.
4. E. R. JETTE and others : *Transactions of the American Institute of Mining and Metallurgical Engineers*, 1934, vol. 111, p. 361.
5. A. J. COOK and F. W. JONES : *Journal of The Iron and Steel Institute*, 1943, No. II, p. 217P.

Radioactive Tracer Technique in a Driving Blast-Furnace

An exploratory trial has now been carried out in a search for reliable data on the velocity of gas through the blast-furnace. No satisfactory results have previously been obtained, and the figures would clearly be of importance in the study of gas flow in the furnace. In this trial use was made of the radioactive tracer technique. A charge of radon gas was projected into one tuyere, and simultaneous sampling of gas from the stockline was commenced. These samples were tested for radioactivity, and a rapid rise in radon concentration was found after about four seconds. This corresponds to much lower velocities than previous estimates. It is proposed to extend the experiment, which is the first successful application of this new technique, to full-scale furnaces in the steel industry.

IRON AND STEELWORKS ENGINEERING

THE IRON AND STEEL ENGINEERS GROUP



THE IRON AND STEEL ENGINEERS GROUP

REPORT OF THE EIGHTH MEETING

THE EIGHTH MEETING OF THE IRON AND STEEL ENGINEERS GROUP, of The Iron and Steel Institute, was held at 4 Grosvenor Gardens, London, S.W.1, on Wednesday, 24th November, 1948. Mr. W. F. CARTWRIGHT (The Steel Company of Wales, Ltd., Port Talbot), Chairman of the Group, occupied the chair.

The MORNING SESSION was devoted to the presentation and discussion of a paper on "Specialized Rolling Stock for Iron and Steel Works," by T. H. STAYMAN and D. R. BROWN, and to the presentation of a paper entitled "Ore Discharging," by G. T. SHOOSMITH. The proceedings of this session will be printed in the August issue of the *Journal*.

The AFTERNOON SESSION was occupied with the presentation and discussion of a paper on "Belt Conveyors, Bunkers, and Chutes for Iron and Steel Works," by J. BRIMELOW, the discussion on which was reported in the April issue of the *Journal*, and by the discussion of Mr. Shoosmith's paper, of which a report is given below.

PROCEEDINGS OF THE AFTERNOON SESSION : 2.0 P.M. to 4.30 P.M.

Discussion on—

ORE DISCHARGING* *by G. T. Shoosmith*

Mr. **G. T. Shoosmith** (Messrs. William Cory and Son, Ltd.), in presenting the paper, said : In this paper I have, in a very brief space, outlined the principal types of equipment used for ore discharging, and brought out what I consider to be some of the important points of design of the man-trolley transporter type.

The man-trolley transporter seems now to be generally accepted as the best type of plant for the average ore-discharging berth, whether it be of the tower type discharging to hoppers, or of the bridge type discharging to hopper and stockpile.

My own firm have used transporters since 1900, and our experience, like that of the Dutch and German firms, is that the slewing-jib type is decidedly preferable when dealing with ocean-going vessels. The peculiar conditions of the Great Lakes have affected American practice, as, obviously, a slewing-jib unit would be largely wasted when discharging Lake vessels, with their narrow-spaced beams.

Dealing with the design side, there are two diametrically opposite schools of thought as regards the structure of transporters or bridges. It can be argued that, since a transporter is subjected to severe and repeated shock in grabbing duty, the structure should be of such mass and of such strength and rigidity that very little deflection takes place, and therefore there would be less likelihood of the riveted joints working. Such an argument means not only higher first cost for the structure but higher cost for the carrying bogies, rails and foundations, or carrying quay.

Against this, it can be argued—and I lean to this view myself—that the structure should be of the lightest possible, that flexibility is often a valuable protection against failure, that welding largely eliminates joint loosening, and that the light-weight structure wins on first cost right down to the foundations.

At the moment, we are experimenting with cast aluminium sheaves, but I cannot yet say whether these will prove successful. As a matter of interest, the aluminium sheave which we use on one crane weighs 138 lb., as compared with the normal cast-steel sheave weighing 296 lb. I look confidently to the wider use in the future of aluminium for both the structural and mechanical parts of transporters and cranes.

There is one opinion expressed in the paper which I should qualify, namely, the matter of single-motor *versus* double-motor hoist-motion drive. Since the paper was written, we have been drawing up designs for some new transporters, which include a solution that I cannot at this stage enlarge upon, but which as far as I know is quite novel. It involves separate drive motors for the hoisting and holding, and will, I believe, give advantages outweighing the electrical complication.

* *Journal of The Iron and Steel Institute*, 1948, vol. 160, pp. 75–84.

I would emphasize, however, that the advantages are mechanical.

There is one aspect of crane design—and here I include transporters—which is, at present, far from satisfactory, namely, the brakes. It is true that the major portion of the slowing down of the load is taken care of electrically, but eventually a mechanical brake must take over, and there is not a straight brake available today which can be applied smoothly, and with any degree of control. I am excluding the air brake, which to my mind is a clumsy, noisy, and expensive way of achieving the objective, and one which does not lend itself to what I regard as the *sine qua non* of single-lever, finger-light control. It is surely not beyond the capability of electrical engineers to design a straight, electrical brake where the speed and power of application are under the driver's control, along with the motor. The ideal brake should be fitted with automatic slack adjustment and should be as nearly noiseless and shockless in action as possible.

On the subject of maintenance, I do not think that I have expressed any views which are controversial.

I should have liked to enlarge on the question of grabs, but this would require a separate paper to do it justice. Great strides have been made in recent years in the scientific approach to grab design, and I am certain that even greater developments lie ahead.

As some indication of what can be done, we recently increased the efficiency of some of our coal grabs by nearly 30%, by simple redesign. The difficulty of establishing grabbing efficiency has been due largely to the fact that individual grab loads cannot be weighed on a normal crane or transporter. Most of our own cranes are, however, fitted with accurate, semi-automatic weighing gear, and this has enabled us to develop a standard test for our grabs, so that we can positively measure the effect of detailed alteration. Hitherto, grab design has been based on 99% of experience and 1% of theory. For instance, there is a widely held view that dropping the grab from a height on to the material improves the digging power, but our experiments have proved beyond doubt that this idea is erroneous.

It may seem odd, in a paper on ore-discharging, to have included remarks on ore-carrying ships. I am confident, however, that anybody who is responsible for an ore-discharging installation will agree that the type of ship is the most important single factor affecting the cost and efficiency of the discharging operation. The example which I have given, of the experience of a particular discharging depot abroad, speaks for itself. This question is of such great importance that I should like to restate the figures given in the paper. If the standard Liberty ship, with a normal rate of discharge of 186 tons/hr., is taken, as being a typical ocean-going freighter, then the discharging rate for a suitably converted ship of this type is 2¼ times as fast, and for a self-trimming ore carrier nearly 6½ times as fast, as normal.

This means that nothing that can be done in the improvement of the design of the discharging gear will have any appreciable effect on the efficiency of discharging, as compared with an improvement in the design of the ship itself. It is obvious, with the present situation of world shipping, that the carriage of all ore cargoes in specially built ships is out of the question for some years to come. My view is, however, that suitable general-purpose cargo ships could be converted to make them, on the one hand, highly efficient from the discharging point of view, and, on the other hand, able to retain nearly all the desirable features for the outward carriage of general cargo.

The proposed plan of conversion of a Liberty ship, shown in Fig. 5, goes beyond anything that has so far been done to this end. This plan of conversion was submitted to one of the classification societies for their comments, and, subject to suitable doubling of certain shell plating to compensate for the cutting back of the between decks, it was agreed that the arrangement proposed would be acceptable.

I would, however, utter a word of warning here. Our experience of bulk-material handling in many parts of the world has shown that the discharging plant must be capable of dealing with the most awkward and unsuitable ships when required.

It is important that we should not slavishly follow American practice, since the ship position in the Great Lakes and in our own ports is entirely different. I have seen, at an American steelworks, grabs being used with a 40-ton gross load, which would hardly enter the hatches of some vessels, and if they did, it would be like a ' bull in a china shop ' when the driver tried to open them. Moreover, in grabbing operations it is not always the largest which is the fastest.

I hope that the few remarks which I have made on the subject of mechanical trimming will not only be of interest, but will stimulate other operators of discharging plant to give their experiences.

The importance of this trimming problem will be appreciated when I say that the labour for hand trimming represents, at many depots, 30% of the all-in cost of discharging, and it can rise to 50–60% when particularly unsuitable vessels have to be dealt with.

Major **W. R. Brown** (Messrs. Ashmore, Benson, Pease and Co., Ltd.) : Not every iron and steel works engineer has the opportunity of gaining experience in ship unloading, and few have any variety of experience. This paper serves a most useful purpose, but I would have liked the author to have been more concise in his recommendations. I am left with the impression that, for our work, he would advocate man-trolley unloaders in preference to jib cranes. I would have liked the author to comment upon some of the conditions in which this sort of equipment has to work, since fog, and high winds have some bearing upon the type of rig which might be chosen. High winds also involve the question of rail clamps which are not mentioned in the paper.

One of the greatest problems in unloading is symbolized in the illustration on p. 81, Fig. 4, in which a massive modern grab is being used side-by-side with the primitive hand shovel. This is an anachronism for which a solution must be found.

Mr. **R. Crooks** (Messrs. Ashmore, Benson, Pease and Co., Ltd.) : I should like to make a few comments on our experience with man-trolley transporters and ore bridges, based on the study of American practice. We have just completed the engineering of three 12½-ton man-trolley ore-bridge unloaders of the American Mead-Morrison design. The man trolleys on these bridges are equipped with a device for twisting a 20-ton (gross) grab so that the grab can be opened longitudinally, transversely, and diagonally to the axis of the ship which is being unloaded, enabling the corners and the sides of the hatch to be reached at all angles. The operator is seated, and controls all the functions with two hand-levers ; with his left hand he controls the movement of the trolley, and also an automatic brake release, which allows the grab to settle in the ore while it is closing ; and with his right hand he controls all the movements of the grab, and has a squeeze-latch with which to control the closing of the grab.

The author gave the reason for the slewing-jib type of man trolley being out of favour in the U.S.A. American designers are very chary of eccentric loading with this type of trolley, particularly on the modern heavy transporter. In view of the fact that the slewing radius

of this type of trolley would normally be of a range to operate within the structure of the bridge or of the transporter in question, we do not think that it would be possible to discharge from one hatch in the ship with two slewing-jib trolleys at the same time. For example, on one pair of rigs operating in this country, the grabs can approach each other to within only about 45 ft., whereas it would seem that the longest hatch on the ship, shown in Fig. 5 of the paper, is about 32 or 33 ft.

There was much diversity of opinion in the U.S.A. on the type of trolley track to be used, but it was not recommended that the rail should be welded to the track beams. We have developed a method of fixing the track rails by special bevel washers fitted over the contour of the rails; and we have flat keepers welded to the trolley beams and fitted up to the rail, preventing any movement in either direction.

Welding, and the use of high-tensile low-alloy steel with high endurance limit, are definite features of the structure of the bridges, and have enabled us to use lighter sections, while retaining stability. The main problem is to reduce the weight of the structure so as to obtain the best possible wheel loads and wheel arrangements on the long-travel trucks, thus minimizing the loads on the foundations. Light-weight construction is not always the answer in this type of structure, which is subject to dynamic loading and where all the parts are moving parts.

Most of the very early man trolleys had a single-hoist motor, with the holding drums connected to it by a clutch, while the opening and closing drums were positively geared to the motor. With the increased grabbing duties of the modern heavy transporter, mechanical engineers demanded two motors, because of the difficulty of designing a compact clutch unit which was reliable, and which would transmit the sudden loads up to, say, a loading of 300 h.p. or more. We do not think that there is much difference in the cost between single-motor-plus-clutch and two-motor drive.

The choice of the correct weight, type, and size of grab, differs very widely in America among both designers and operators. Most of the large outfits are equipped with large clean-up grabs of 7 to 8 tons capacity, with a spread of up to 24 ft. when open, which are used after the regular grabs have reached the limits of the hatchways.

High-tensile steel is used extensively in the construction of coal grabs, as the decrease in dead-weight of the grab can be used to increase the pay load. On ore grabs this weight is actually required to ensure penetration of the jaws into the ore while closing. We find that the weight of the grab is a definite factor. It is not so much a question of the height from which the grab is dropped into the ore, but of the actual weight of the grab when it has settled on the ore, that prevents it from closing without lifting its full capacity.

Mr. **J. Russell Taylor** (Igranic Electric Co., Ltd.): In regard to the use of D.C. or A.C., there are cases where A.C. is cheaper in the long run, and other cases where D.C. works out almost as cheap, or even slightly cheaper; but it is generally recognized that with D.C. it is possible to get much better control over slow-speed operation. When a ship is being emptied and the bottom is reached, it may be necessary to lower a large grab, closed through the hold, and then open it, swing it into a corner and rotate it on its own axis, so as to place it exactly where it is required, and then close it gently, to scrape the bottom. In such a case, D.C. is more flexible than A.C., and the two-motor arrangement, with two separate motors electrically controlled by two separate levers, gives greater control of the grab.

Dr. **L. Bramley** (British Iron and Steel Research Association): I rise primarily, and not for the first time, to plead the cause of the smaller master controller, and the improvement in cab design which a smaller controller will permit. The author has stated the case for light, and well arranged controls, but the control handles in Fig. 3 neither look finger-light, nor appear to be in the best positions for continual use. The driver should be seated, and his movements should be from the wrist or elbow rather than from the shoulder.

It would be interesting if the author would enlarge on the statement which he makes concerning the face-plate type of controller, because at B.I.S.R.A. we are starting to look into the possibilities of building a small master controller which might be of the face-plate type.

The author mentioned cabling in heavy-gauge galvanized conduit. I have seen many cranes equipped with such protection for the cables, and it is my experience that it is inadequate. The joints fail at the corners, chiefly because of fatigue, and altogether it will not give a long life. I would advocate steel trunking $\frac{1}{8}$ in. thick and 3 to 4 in. square, with the movable lids and fastenings arranged so that they cannot be detached and lost.

Mr. **W. J. Pool** (The British Thomson-Houston Co. Ltd.): When it was decided, a year or two ago, to install Ward-Leonard control on some ore bridges in this country we were told that it was contrary to American practice, which was to utilize constant voltage D.C. However, on a recent visit to the U.S.A., I found that Ward-Leonard control was being seriously considered, and was almost certain to be adopted, for some very large, new ore bridges then in the proposition stage. This change from constant voltage D.C. to Ward-Leonard control follows the lead already set in the steel mills, where the most heavily worked auxiliaries are now almost invariably put on Ward-Leonard control. Constant voltage D.C. or A.C. supply, used directly on the motor of such an equipment, inevitably means heavy control gear, and I have always understood that one of the advantages of Ward-Leonard control for this type of work was reduced maintenance. I am surprised, therefore, to find that the author puts increased maintenance as a disadvantage of Ward-Leonard control.

From the operating point of view, Ward-Leonard control has the advantage that the motor inherently does, at least approximately, what it is instructed to do through the control, *i.e.*, whether the hook is light or heavy, and whether the load is being raised or lowered, the direction, and within rough limits, the speed, depend primarily on the controller position. On the other hand, if an A.C. supply is used directly on the motor, and this applies equally to constant voltage D.C. with the simplest form of control, the direction of movement and speed of the hook depend primarily on the load, and only secondarily on the state of the control. This is avoided on straight D.C. by potentiometer control, at the expense of a rather heavy power consumption. In the case of straight A.C. a number of ingenious ways have been, and are still being, brought out for overcoming the difficulty, but they all involve complications, and those that approach most nearly to the results obtainable with D.C. usually also involve increased heating. In fact, where it is necessary to have anything like accurate speed control, it is very difficult to equal Ward-Leonard control.

However, many grabbing-crane installations are an exception to this requirement of accurate speed control, as in the case of just grabbing off a pile. In such cases, sufficient control can very often be obtained by simple means with straight A.C. motors. I can quote a case of a two-motor grabbing crane where two slip-

ring motors are used quite successfully with perfectly standard and straightforward reversing control, the only special feature being that a common joy-stick, working in a gate, is used.

Some years ago I considered various alternative methods of operating grabs with straight A.C. motors, and came to the conclusion that this was a case where a little complication on the mechanical side might save a great deal of complication on the electrical side. There appeared, as Mr. Shoosmith has said, to be a great deal in favour of using a single motor and a mechanical clutch. I should be interested to know what has led him to depart from this scheme in his latest installation. In the single motor and clutch installation, described in the paper, was any arrangement made to prevent the clutch being engaged when the motors were at speed ? If this is not done, there is risk of considerable wear and tear. Were arrangements made to release the holding motor brake independently, to allow the grab to settle into the pile during digging ? Is the motor actually totally enclosed, or is it forced-ventilated ? Is it continuously rated ? This would appear to me to be correct. A truly totally enclosed, continuously rated motor, however, is apt to be excessively bulky and to have too high an inertia.

The author mentions the desirability of having control by an electro-mechanical brake, by which the braking effort can be gradually applied and controlled. This can be done, but at the expense of a certain amount of complication. One solution which occurs to me, is to use a spring-loaded brake, torque-motor released, the torque motor being a slip-ring machine with adjustable secondary resistances. The same effect can be produced, though rather less satisfactorily, with a squirrel-cage motor and adjustable primary resistances.

The author has mentioned the use of a common joy-stick control for the hoisting and slewing motions. I believe that this is a fairly common practice, but I should have thought that it was more logical to use a common joy-stick for slewing and luffing, so that a single movement of the lever in the horizontal plane controlled the position of the grab in the same plane. I wonder why the particular combination referred to by the author has been chosen.

Can the author say anything about his experiences on the travel motion of grabbing cranes ? This type of crane usually has a base plan which is approximately square, and one would assume that the structure is sufficiently rigid to dictate the relative speeds of the motors, of which there are often four, one on each leg. If this is the case, and the wheels are of slightly different diameters, or the crane is travelling on a slight curve, the mechanical conditions would compel the motors to run at different speeds. With A.C. motors, an obvious way of permitting this, and one which is, I believe, generally used in the U.S.A., is to employ independent rotor resistances with mechanically connected contactors, or a common master control. In this country it is sometimes specified, in similar cases, that the motors should have a common rotor resistance. This resistance can only be effective, however, if the motors are truly synchronized, and they can only synchronize themselves by making use of backlash and spring in the crane structure. In the case of a crane having four travel motors on a D.C. supply, the simplest and cheapest way is to put them all in series and use a common controller. They are then perfectly free to accommodate their relative speeds to the requirements. This arrangement has the one possible defect that, if the structure is so rigid that it is possible for the wheels driven by one motor to lose their adhesion, this particular motor will run away and the others will stall. Whether this is a real danger or not appears to depend on the mechanical design of the crane structure.

Mr. **G. W. Grossmith** (Strachan and Henshaw, Ltd.) : In various parts of the world, under storm conditions, transporters have been driven along their tracks to disaster, so the question of securing them is important. Rail grips, hand operated or automatic in their action, are effective in varying degrees, but frequently it is inconvenient to have the rails projecting above the general level of the quayside, and to provide grooves along both sides of the rail is not a practical arrangement, as such grooves become clogged. The question arises, can a transporter be secured against storm conditions by any other method than by gripping the rail ? I have made calculations for a number of transporters of varying sizes and proportions, under an assumed wind pressure of 35 lb./sq. ft. of exposed surface, and have found that in no case would the transporter move, provided all the wheels were locked. In each case, also, the machine was stable against overturning in such a storm. Most transporters are driven by two bogies out of a total of four, so that to hold all the wheels requires that the wheels of the undriven bogies should be locked, with brakes holding the wheels of the driving bogies, which is what we are now doing.

Referring to the hoist-motion drive, one advantage of separate motors, as compared with a single motor and clutch, is that, although there is little difference in the total number of mechanical parts in each case, the twin drive system is mechanically simpler, and I prefer simplicity in duplicate to complication solo.

Mr. **R. A. West** (Igranic Electric Co., Ltd.) : I should like to ask the author a few questions about cross-traverse brakes. When he refers to brakes being gradually applied, he is probably referring to the cross-traverse drive of the man trolley. It is desirable for this braking system to be graduated, but I doubt whether the extra complication of making an electric brake act gradually is really worth while, or whether it is desirable to supersede the driver's function of applying the brakes gradually.

There are several ways in which the operator may be provided with graduated control of braking. With D.C. there can be graduated dynamic braking, or plugging, which is more commonly employed and which can also be used with A.C. If there are two or three steps, there is a considerable choice of braking effort sufficient for the cross-traverse motion. Those systems are fairly simply obtained, and, although the author puts a question to the electrical industry, I should like to hear him amplify his remarks as to the need for a gradually applied electro-mechanical brake.

Mr. Grossmith mentioned rail clamps, and I should like to ask whether they are automatic. It would seem to require considerable care in instructions to the operating staff to see that the clamps are inserted. We have been concerned with many outdoor structures on which storm brakes are automatically applied in different ways, but the only safe one seems to be the type which has a weight which falls by gravity when the power is switched off, and actually pinches the sides of the rail.* That does not overcome Mr. Grossmith's point of what to do when there is no rail to grip ; will Mr. Grossmith comment on the question of whether the operator might forget to put in the storm brakes when he leaves the structure ?

Mr. **E. T. Judge** (Messrs. Dorman, Long and Co. Ltd.) : We are not interested in unloading rates of 1000 tons/hr.,

* The weight is lifted by means of a miniature winch ; rheostatic braking for the motor of the winch in the lowering direction allows the weight to fall slowly.

but in the minimum cost of getting a complete ship unloaded. In this connection, the difference between conditions here and the conditions in the U.S.A., in regard to the type of ship that is to be dealt with, has a big influence on the type of equipment most suited to our conditions.

The author drew attention to the high cost of hand trimming. If labour can be provided for hand trimming only when it is required, then probably the cost of getting out the last 20 to 25% of the ore is somewhat more than the cost of getting out the first 75 to 80%. If you consider the installation of a somewhat larger unloader than normal, which might reduce the cost per ton of breaking down by, say, 10%, it is possible that the more cumbersome equipment would necessitate hand trimming after dealing with, say, 75% of the ore, whereas a somewhat smaller tool might get out 80% from the ship. In that event, 5% more left in the ship would be such a burden to the cost that, despite the cheaper breaking-down cost, the total cost of unloading the ship would increase. Therefore, the first point which needs emphasis in regard to unloaders, for the ships with which we have to deal, is how flexible a machine can we install to grab, without trimming, the maximum amount from the general run of ships. With this consideration in mind, my own company are installing swinging-jib-transporter unloaders, limiting the gross weight of the unloader to 19 tons, and we are putting in level-cut grabs with D.C. control. We hope in that way to get the maximum out of the ship before trimming.

A problem in this country is to provide trimming labour when it is required, and to do without it when it is not required. Are we right in accepting a very expensive tool, with a 7, 10, or 15-ton grab on it, with men in the hold shovelling material to it so that the material can be lifted out ? In our case, we are fortunate in wanting to unload from ships, under normal conditions, between 30,000 and 40,000 tons/week, and at the site where our new plant was to be installed, we had available an existing berth equipped with four jib-cranes. We intend to use two unloaders on a new berth to do all the breaking down, so far as we can work with the grab alone with no men in the hold, and then to move the ship up to the existing berth, and remove the remaining 20%, or less, by means of the four jib-cranes. We believe that our two unloaders breaking down will be able, under normal conditions, to keep ahead of the four jib-cranes which are tubbing-out hold bottoms.

We are concerned with ways and means of improving the removal of the last 20% from an ore ship, and I wonder whether the use of comparatively cheap jib-cranes, with, if necessary, a modified form of tub, cannot be mechanized more easily than sweeping the material from the hold corners, and using expensive grabbing equipment to lift it out. We are now installing special iron ore loading-machines in our ironstone mines, which are compact, for working underground, and which effectively fill tubs, which are on wheels in this case. If this equipment could be modified to mechanically fill tubs in the hold of a ship, with jib-cranes to bring the tubs out, the last 20 to 25% of ore might be much more cheaply removed than by trimming to a grab and having expensive unloading equipment, which can run at the rate of 600 tons/hr., or more, tied up in lifting the dregs out at a rate of less than 50 tons/hr.

The **Chairman** : Everybody who has worked Ward-Leonard control knows how ' sweet ' it is, but it seems ridiculous to put in this large mass of rotating machinery. I was recently privileged to see a D.C. 200-h.p. motor controlled by an induction regulator and mercury-arc rectifier, which seemed to work as sweetly as any Ward-Leonard control, to be quite noiseless, to occupy very little space, and to weigh very little. Have we reached the point when, instead of carrying a rotating Ward-Leonard control on the crab of the ore bridge, taking up a lot of space and making a noise, we can put a saturated core reactor, as the Americans call it, or an induction regulator to vary the input voltage to a mercury-arc rectifier, and thus get a Ward-Leonard effect ?

I understand that the old objection was that you could not pump back when you wanted to brake, but a form of valve can now be fitted which makes it possible to pump back in braking. Is this equipment sufficiently advanced to enable people who want to buy ore unloaders to put in a metal-clad pumpless rectifier and other equipment, as robustly designed as is the present Ward-Leonard plant ?

I should also like to ask the author, who has much more experience of ships than most of us, why, when a tanker can be built up to 30,000 tons with the engines aft, people persist in building ships, like the Liberty ships, with the engine in the middle and a tunnel-shaft aft, which gets into the teeth of the grab.

Mr. **H. S. Carnegie** (English Electric Co., Ltd.) : Referring to the Chairman's question, the use of rectifiers instead of a motor-generator set for Ward-Leonard crane work is quite possible, but is not practicable in all cases. It is obvious that the rectifier possesses advantages in efficiency, no load losses, and noiselessness, but has the great disadvantage of being uni-directional. This latter fact makes it much more difficult to deal, not only with reversing loads, but also with overhauling loads, as in the case of crane or similar mechanisms. In due course, no doubt, these disadvantages will be surmounted, but in the meantime, the use of rectifiers in place of Ward-Leonard control for duty similar to that of a crane is not so very practicable.

The author refers, on page 79, to first cost, consumption, and maintenance cost, of crane motors. I would like to query the apparent placing of first cost in a position of primary importance. Surely that position should be taken by the cost per ton of material shifted, taking into account the reliability, maintenance, period out of service, etc. over a term of years ?

The author mentions the use of motors of 1000 r.p.m. I would suggest that this speed is rather high and would increase the inertia of the motor and of the total drive, to which the author has so rightly called attention.

The relative advantages of various A.C. and D.C. systems for cranes must undoubtedly depend on the size, frequency of operation, and manœuvrability required. Owing to the simplicity in equipping, and also in feeding, an A.C. crane, it will undoubtedly be the first choice for cranes used infrequently, but there will be more advantage in using D.C. when greater manœuvrability or greater frequency of operation is required. When considerable flexibility is needed, the Ward-Leonard control offers many advantages. However, Ward-Leonard gear introduces rotating converters with the consequent increase in noise, and this, no doubt, has suggested the Chairman's mention of rectifiers.

The author mentions D.C. motors as having higher inertia than A.C. machines, but, in general, the opposite is the case, particularly when the speed is low or the power is increased. Moreover, D.C. mill motors are so standard in steelworks that they obviously should be suggested for important jobs. There is then a minimum of inertia, the machines having about half the inertia of industrial A.C. motors, which will affect the total inertia of the hoist mechanism by about 20%, so decreasing the inertia torque by that amount.

CORRESPONDENCE

Mr. **J. Baker** (Messrs. Clyde Crane and Booth, Ltd.) wrote : From a crane maker's point of view the paper is very welcome. It is particularly pleasing to find that a user, of Mr. Shoosmith's experience, is prepared to discard preconceived ideas and fashions covering the practice of ore unloading, and to deal with each particular method on its merits. He recognizes the fact that excessive use of material is expensive, and does not necessarily mean satisfaction from a working point of view. This attitude is welcome at this time when there is a tendency to push weight into structures, with the idea that it is playing for safety.

In the case of the structures the word 'stable' rather than 'rigid' is used. In many cases flexible structures have withstood loads which would have caused trouble in rigid structures. Similar trouble was experienced in the early days of welding, when designers failed to recognize that a welded joint was rigid compared with a riveted joint, and that equivalent flexure must be arranged for in the structure itself to replace that which previously took place at the joint. Whilst this fact is recognized, I have not yet been able to apply technical reasoning to it, or suggest methods of calculating what allowances should be made. I think, however, that methods of calculating structures such as ore bridges can be vastly improved upon, and with the modern technique in use for working out the stresses in steel structures, we shall be able ultimately to design ore bridges lighter than those at present in use, but with a sufficient margin of safety.

In the case of a man-trolley type of transporter the weight of the trolley is of vital importance, both from the point of view of overall weight of the structure, and from the point of view of cutting down inertia in order to obtain high rates of acceleration and braking. I agree with Mr. Shoosmith that springing of the trolleys is of no benefit; springs which are heavy enough to carry the high axle loads involved, have a natural frequency far too high to be of service in damping out vibration due to rail joints. Mr. Shoosmith mentions rail joints made continuous by welding, but I have not found that this is the answer to the problem, and if he has any information where welded joints have withstood continuous heavy wheel-loads (as opposed to railway wheel-loads, which are light compared with crane practice) I should be pleased to have further particulars.

Although high acceleration and braking are important features in obtaining a high cycle of output, I would also emphasize that it is important that there should be some degree of control of the rate of acceleration and braking. This is necessary when closing the grab, so that the speed of closing can be eased off just as the closing operation finishes, and the grab can be picked up on the holding rope without jerks. It is also particularly necessary on the traverse motion : as soon as the trolley starts to move, the grab tends to lag behind ; the trolley must then be slowed down until the grab starts its forward swing, when the trolley can be accelerated up to full speed, thus damping out the swing of the grab. Without this degree of control much time is lost.

It is refreshing to find Mr. Shoosmith an advocate of A.C. for transporter drives. This is one of the lines of argument which I took up at the Meeting some months ago, when A.C. *versus* D.C. supply for cranes was discussed. At that time there was an almost overwhelming majority in favour of D.C. I do not agree, however, that A.C. is necessarily cheaper in first cost; I think there is very little difference between the two. The chief argument against A.C. from a crane maker's point of view is the difficulty in accommodating the extra collector bars required. From the output point of view, D.C. has no advantage over A.C., and I fully endorse Mr. Shoosmith's remarks to the effect that there is no need for excessive maximum speeds, either of traversing or of lowering.

With hoisting and traversing motors, which in some cases have a higher R.M.S. rating than the maximum continuous horse-power rating, because of the frequency of starting and stopping, I would consider it an advantage, from the point of view of weight saving, to use forced-draught ventilated machines. The only objection to this practice is the noise made by the ventilating fans, which is even more objectionable than the hum of a Ward-Leonard set. A point which Mr. Shoosmith does not make in regard to the mill-type motor, is that it is usually designed with a low-inertia armature, and although I agree that there is no call for a mill-type machine, where R.M.S. rating is the criterion of motor size, care should be taken to select an industrial design of motor with a low-inertia armature or rotor.

The question of the number of motors for hoisting is interesting, and, apart from the merits of one-motor drive *versus* two, resolves itself into a question of accommodating the equipment within limited weight and space dimensions. The two-motor arrangement is a definite advantage in this direction, and one might even consider a four-motor drive, *i.e.*, two motors to each drum, in order to secure economical use of space. A further advantage of this latter arrangement is that the combined inertia of four rotors is very much less than that of one large machine, and it is also possible to use four brakes to spread the heat over a larger brake shoe and drum area.

AUTHOR'S REPLY

Mr. **Shoosmith** in reply said : Major Brown referred to the importance of the driver's view on foggy days, and to that I would add windy conditions when dealing with a very dry material, when there is a large amount of dust in the hold. We think that this question of the driver's view is very important indeed, and it is one of the reasons why we favour the slewing-jib type of transporter rather than the grab-twisting type, inasmuch as it enables the driver to see his grab, even when working under the near-side coamings, which he cannot do with the grab-twisting type.

I also agree about the importance of rail clamps. American practice has generally been to use magnetic rail clamps ; they are enormous devices, sometimes with an anemometer for overriding control, which automatically applies the clamps if the wind velocity rises to a dangerous point. Our experience here is that the ordinary type of rail clamp, such as is fitted to standard cranes, is useless ; with a wind sufficient to shift the crane, the clamps will open. We substitute a scissors type of clamp, a simple mechanical device, easily applied by one man, which has the advantage that the harder the crane pulls on the clamp the tighter the clamp grips the rail. We have had no trouble with these.

I have given a general picture of the types of plant which can be used, with the idea of emphasizing that each case must be considered on its own merits. When dealing with a normal steelworks problem, in general I think it will be found that the man-trolley type is at the present time the most efficient machine.

In reply to Mr. Crooks, on the question of grab twisting *versus* grab slewing, I do not agree that a grab-twisting device can be worked with the same precision as can a slewing-jib type. Moreover, with the slewing-jib type, a skilled driver—and it is extraordinary how skilful these men become—can manœuvre his crab and his slewing gear in such a way that he can throw the grab

right underneath the ship's coamings. At first sight it looks dangerous, but a skilled driver can almost drop the grab on to a nut in the hold, which emphasizes the importance of the controls being absolutely finger-light, easy to handle, and positive in action.

On the question of two grabs working in one hold, the American type of transporter is usually fitted with very wide-spaced carrying-bogies for the long travel. I do not know why that is, unless it is that everything is a little bigger and wider over there ; they seem to have more room to play with. In working out the wind pressures on the transporter structure, it is generally found unnecessary to place the long-travel bogies at anything like the centres often used. With some machines which we are designing at the moment, we have our outside, long-travel bogie wheels within the main uprights of the structure, with the result that with two transporters working alongside, with the two jibs of those two transporters swung round towards each other, there is a gap of about 15 ft. between the centre line of the two grabs, so that it is quite possible for those two grabs to work out of the same hold.

The question of the traverse rails is one about which I have not made up my mind. I was interested to note in the U.S.A. last year that coned traverse wheels, and traverse rails with a cant, were being used, with the object of equalizing, and presumably, improving, the grip on the rail, and to give better wear, instead of more or less point contact. I do not know from experience whether that provides any advantage over using a flat rail and a wheel with twin flanges.

On the question of two motors or one, in presenting the paper I have had, to some extent, ' to eat my own words.' I agree that the question is mechanical rather than electrical. We have not found that the clutch mechanism is really the limitation of the one-motor drive ; we have made clutches work with extraordinary precision, and with a long life, even in the largest sizes. The gearing really affects the question, and may lead to a decision in favour of multi-motor drive.

I do not agree with Mr. Crooks about the clean-up grab. In a number of plants that I visited in the U.S.A. I noticed clean-up grabs rusting in a corner of the yard, and was told that the operators never used them, because it was too much trouble to change them over. I favour the use of a dig-and-scrape type of grab or a Voorwinde type, which is a wide-spread grab without articulated pans, which can be used right through the ship, to avoid the trouble of changing over a grab when getting down to the ceiling of the ship. Grab efficiency is largely a question of weight, but the spread of the grab also enters into it (particularly when it is a matter of digging difficult material such as Swedish ore or large, lumpy material), and also the closing purchase. This is an elaborate subject from a theoretical point of view.

In reply to Mr. Russell Taylor, I must agree that the D.C. drive, particularly the Ward-Leonard type, is more flexible and somewhat smoother than anything else that I have seen to date, but experience shows that we do not need that additional last degree of smoothness and flexibility ; the control which we can get with A.C. is wholly suitable for our particular grabbing duty, which is usually out of a ship.

I mentioned in the paper that when our latest A.C. cranes were erected, I sent down from another depot, where we have Ward-Leonard cranes, one of the best drivers, who had spent most of his life driving Ward-Leonard machines. After about ten days on the new A.C. machines he told me, without any prompting, that he liked them better than the Ward-Leonard cranes, partly because they were later machines, but he said that the control was perfectly satisfactory from his point of view, and in every way as good as our Ward-Leonard machines.

In reply to Dr. Bramley, I think that perhaps the photograph of the driver's cab (Fig. 3) in the paper, is not very clear. The cab is very small, and these controllers, which Mr. Russell Taylor built, are face-plate controllers and can be moved with two fingers, by a touch. They are absolutely finger-light, and there is a fair amount of movement on them. We like a fair amount of movement on the control lever, because in our view it gives better control.

As far as conduit is concerned, we have always found extra heavy galvanized conduit with gas thread, not electrical thread, very satisfactory. Possibly the very heavy built-up conduits, which were suggested, would be better, but they would be much more expensive.

Mr. Pool also referred to the Ward-Leonard drive. The disadvantages, of course, are the extra weight, bulk, and noise. I have not yet met a Ward-Leonard set which had not a high-pitch whine of some kind, and that can be very disturbing to a driver for eight hours a day.

On the question of maintenance, with the Ward-Leonard machine there are, in effect, twice as many machines to look after, and the brush-gear is a good deal more troublesome than the brush-gear on an A.C. machine. On the question of power consumption, the figures quoted in the paper were taken from a D.C. Ward-Leonard depot, and I gave an average figure of 2·5 tons/kWh., as against a figure for the A.C. depot of 3·5. Over the last twelve months our A.C. depot has averaged 3·55 tons/kWh., and that includes all the lighting, of both the berth and the offices, and the power consumption of a small workshop, and of a small wagon tip.

On the question of digging into difficult material, Fig. 3 shows the dig button, marked (4). That button momentarily releases the holding brake to enable the grab to sink into the material. Normally, that dig button is used only with particularly difficult material.

With regard to the engagement of the clutch, on our latest cranes, where we have single-motor drive, we have arranged that, until the clutch is engaged, only the first notch on the controller can be operated. Wherever the driver puts the controller, the electrical equipment only gets into the first notch, so the grab closes comparatively gently. As the driver releases his clutch button, the controller goes up to the full power corresponding to the position of the control lever. With regard to motor rating, as far as hoists are concerned our motors are continously rated on R.M.S. value, taking the heaviest possible duty cycle (breaking down cargo).

I have mentioned the working of the control levers. I was asked why we put hoisting and slewing on to one control, instead of hoisting and luffing. The answer is that our levers move in the same direction as the load ; if you put the lever down you lower the load, and if you raise the lever you raise the load, similarly with movements to the left and right.

On the question of the long-travel electrical control, I do not want to enter into a long discussion, especially as we have had practically no trouble with long travel, probably because our transporters and slewing cranes and other equipment do not use this long travel very much, unlike an overhead traveller on which the long travel is continually in operation. We have various types of long-travel drive—mechanical cross-connected, D.C. individual motors, A.C. individual motors, etc., and they all work satisfactorily

In reply to Mr. Grossmith, I have mentioned the importance of the rail clamps, and I agree that it is

difficult to provide proper clamp arrangements, or safety brakes, where the rails are sunk, and it is, of course, an enormous advantage to sink the rails to have a flush top on the jetty or unloading berth. I also agree that if all the wheels can be locked, in ninety-nine cases out of a hundred, everything will be all right. I also agree on the advantage of duplication of the motors as far as the multi-motor drive for the hoist motion is concerned, and especially if it is possible, by so doing, to use the same motors, and possibly some of the same gearing, as on other motions on the crane. That is a point which we are looking into at the moment.

In reply to Mr. West, in introducing the paper, my remarks about brakes did not apply to the cross-traverse, because that can be stopped by plugging, and once stopped it will remain stopped, unless there is a particular swing on the grab, or a very strong wind. I doubt, even then, whether the crab would move. It is not so important on the cross-traverse as it is on the hoist, where after the load is plugged to a creep it may be still suspended. That is when a mechanical brake is needed which will come in smoothly, but positively, and under the control of the driver or under some variable control, so that it can be set to engage the brake smoothly without jarring the whole machinery when the load is suddenly stopped.

In reply to Mr. Judge, I agree that the real problem is the overall cost of discharge, and not the cost of discharging the cream of the cargo. I also agree that too large an equipment may be more of a disadvantage than an advantage for the type of ship with which we have to deal. I saw an example of that in the U.S.A., where at one discharging berth, which had to deal with ocean-going ships, it was found that 35-ton (gross) machines could not be used on ocean freighters of the normal type, and the whole of the discharging had to be done with small 15-ton equipment. In other words, their best plant was lying useless at that time. However, I cannot agree with Mr. Judge on the question of moving the ship. This is an admission of defeat, and in our experience it has all sorts of complications. Sometimes masters will not move ; they say that there is a cross-wind or fog, or that the crew are not there, and the ship may be in berth whether it is wanted there or not. I think that the suggestion of trying to get out the tail end of the cargo with tubs is retrogressive. In our experience it is expensive in labour, and even more so if the plant has to be duplicated. We have seriously thought of trying out a type of equipment, such as the Joy loader, for cleaning up the hold, which, I imagine, is the type Mr. Judge has in mind. Such equipment is extremely expensive and heavy ; in fact, so heavy that our normal cranes would not be able to handle it, if it were of sufficient capacity for the job. In our view, it is better to do the whole job at one berth, under one set of cranes, and to tackle the trimming mechanically, which I am certain can be done.

In reply to the Chairman, who asked the reason for central-engine ships, I am afraid I do not know the reason for it. My father has told me of the arguments that used to go on in the old days between the steamship side (the operating side in regard to colliers) and the engineering side, the former wanting to have the engines amidships, and the latter wanting to have them aft, for ease of discharging. It was said that the ships would be unsafe, and would break their backs, if the engines were aft, and it took years to break down that prejudice, but eventually it was done by building colliers with the engines aft, purely for short coastal trips, and then, I imagine, some emergency arose and they had to be sent somewhere abroad. Today, nobody would dream of building a collier with the engines amidships.

Tankers are a class of their own, because they have immense longitudinal strength due to the nature of their construction ; but with a collier, special steps have to be taken to provide the requisite strength longitudinally, and, incidentally, hopper tanks and top-side balance tanks help to do that. Whether it is prejudice which keeps the engines amidships in the ordinary seagoing freighter I should not like to say, but I think that has much to do with it.

During the war we had sent over for use in this country a number of Canadian lake vessels, which, in certain respects, are even better from a discharge point of view than the colliers, because not only are the engines aft but the accommodation is right forward.

In reply to Mr. Carnegie, I agree that first cost is by no means the main consideration ; it is the overall long-period cost which counts. I still maintain that on the overall long-period cost basis A.C. will win. I do not agree that D.C. is any better than A.C., even for continuous duty. As a matter of interest, our three new 7½-ton gross load A.C. cranes discharged, last year, well over 1,000,000 tons, and that is fairly continuous going. I do not think that we had more than a total of 5 minutes' hold-up for the electrical equipment during the whole of that time.

REPLY TO CORRESPONDENCE

Mr. **Shoosmith** wrote : In reply to Mr. Baker, I would mention that we have in many instances welded the joints of the longitudinal carrying rails, with every success. We have found it necessary, however, also to weld fishplates on to the joint so as to give additional strength. In the same way, and in order to avoid jar, we have welded the butts of sectional race-paths on slewing cranes. I fully agree with Mr. Baker's remarks regarding the economy of space which can be achieved with a multi-motor hoist-drive and with careful layout. I agree also that such an arrangement is advantageous from the point of view of low inertia and larger braking surface. I am glad to have Mr. Baker's support for the use of A.C. drives for transporters. In my experience the adoption of A.C. has so far resulted in reduction in first cost, although, admittedly, the reduction has not been very great. The real benefit of A.C. drive, however, is in its lower maintenance cost, and our experience has been that this reduction is appreciable.

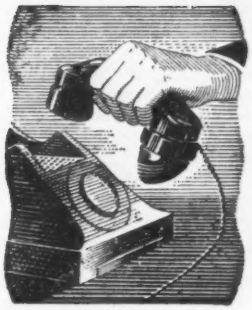
A Unified Thread Form

A provisional specification for the new thread series is being issued by the British Standards Institution, which will be available for use by industry, and will be adopted eventually by Government Departments.

The immediate impulse for unification of British and American screw-thread practice came from interchangeability difficulties during the late war, and it will be appreciated that such a unification will be of great value to Anglo-American trade in peacetime.

It has been found that, in general, the new form is somewhat stronger than the Whitworth thread, though the difference is only marked in cold-rolled threads. Because of the high degree of interchangeability between the B.S.W. and the Unified thread forms, combinations are likely to occur of nuts of one form on bolts of the other, during the period following the introduction of the new form. Tests show that no loss of strength is occasioned by such "crossing," except, possibly, in the case of high-tensile nuts, the use of which is, in any case, undesirable from the aspect of strength.

N E W S



ANNOUNCEMENTS AND NEWS OF SCIENCE AND INDUSTRY

THE IRON AND STEEL INSTITUTE

Changes on the Council

Mr. GUNNAR SCHJELDERUP, Managing Director of Christiania Spigerverk, has been nominated to be an Honorary Vice-President, in recognition of his services as Chairman of the Reception Committee in connection with the Special Summer Meeting of the Institute, in Norway.

Corrosion of Steel

A Joint Meeting of The Iron and Steel Institute and the British Iron and Steel Research Association, to discuss several papers dealing with corrosion and corrosion-fatigue of steel, will be held at 4 Grosvenor Gardens, London, S.W.1, on Tuesday, 18th October, 1949. There will be two sessions, at 10.30 A.M. and at 2.0 P.M. Further details will be given in future issues of the *Journal*.

Autumn General Meeting, 1949

The Autumn Meeting is to be held in London on Thursday and Friday, 10th and 11th November, 1949, and not on 9th and 10th November as previously notified.

Fourth Hatfield Memorial Lecture

Sir ANDREW MCCANCE, LL.D., D.Sc., F.R.S., President, has agreed to deliver the Fourth Hatfield Memorial Lecture, in Sheffield, on Friday, 28th October, 1949. His subject will be, "The Plastic Behaviour of Solids."

Symposium on High Temperature Steels and Alloys for Gas Turbines

The Council have made provisional arrangements for a Symposium to be held on Thursday and Friday, 19th and 20th October, 1950, on the subject of steels and alloys for use under high temperature conditions, such as in gas turbines.

The Fifth Hatfield Memorial Lecture will be delivered, on the Wednesday evening prior to the Symposium, by Sir Frank Whittle, K.B.E., C.B., D.Sc., F.R.S.

Relations with Foreign Societies

The centenary of the birth of Floris Osmond is to be celebrated at a Meeting of the Société Française de Métallurgie, to be held in Paris during the week commencing 3rd October, 1949.

Dr. C. H. DESCH, F.R.S., has agreed to accept an invitation to join the Committee of Honour, as a representative of The Iron and Steel Institute.

The Autumn Meeting of the Institute of Metals will be held in Paris, in conjunction with the Society's Meeting.

Special Reports

Three publications in the Special Report Series are to be issued in the course of a few weeks. Brief particulars of each report are given below, and applications for copies from Members and others are invited. Requests, with remittances, should be addressed to the Secretary, 4 Grosvenor Gardens, London, S.W.1.

Atlas of Isothermal Transformation Diagrams—Special Report No. 40 (Price 25*s*., Members, 15*s*.).

Prepared by a Group of the Thermal Treatment Sub-Committee of the Metallurgy (General) Division of the British Iron and Steel Research Association, this report comprises isothermal transformation diagrams of 24 steels in the British Standards Institution EN Series. Each diagram is accompanied by details of the tests, supplementary isothermal data, and comments relating to steels within the EN specification range.

Corrosion of Iron and Steel by Industrial Waters and Its Prevention—Special Report No. 41 (Price 5*s*., Members, 3*s*. 9*d*.).

This brochure has been prepared by the Industrial Waters (Corrosion) Sub-Committee of the Metallurgy (General) Division of the British Iron and Steel Research Association. It surveys, in a general manner, existing knowledge of the corrosion of iron and steel by industrial waters, and of the preventive measures which might be employed. It is illustrated by typical examples of corrosive action.

Report on the Bessemer Process—Special Report No. 42 (Price 25*s*., Members, 15*s*.).

The report reviews existing and past Bessemer practice in this country and abroad. It enumerates the uses to which Bessemer steel is put, and indicates possible lines of research which might lead to the development of new applications for Bessemer Steel.

The main sections of the report deal with British Bessemer Practice, Foreign Bessemer Practice, and Future Bessemer Practice; details of particular plants are given in appendices. The report has been prepared by the Ad Hoc Bessemer Sub-Committee of the Steelmaking Division of the British Iron and Steel Research Association.

Adoption of the name "Celsius" in place of "Centigrade"

When the Metric Convention was signed in 1885, an International Bureau of Weights and Measures was set up. Meetings of the signatories of the Convention are held at periods of six years, to review the work of the Bureau.

At the Ninth General Conference on Weights and Measures, held in Paris in October, 1948, the name of the Centigrade temperature scale was considered. The temperature scale, in which the interval between the ice and steam points is divided into a hundred parts, is known in this country and in France as the Centigrade

scale, but the use of this word in French leads to an ambiguity, since the "grade" is the French unit of angular measure in which the right angle is divided into a hundred grades. In consequence, the use of "Centesimale" has been suggested in France. In Scandinavia and Central Europe, the scale is known by the name of its originator, Celsius. The Ninth General Conference agreed that the same name for the scale should be used in all countries, and decided in favour of Celsius. This plan has the advantage of not changing the abbreviation "°C.", which is common to all languages, and it also has the effect of naming all the temperatures scales in the same way, each being known by the name of its originator: Celsius, Fahrenheit, Kelvin, Réaumur.

In common with other societies, and the British Standards Institution, the Journal of The Iron and Steel Institute will, in future, adopt the new term.

NEWS OF MEMBERS

▶ Mr. R. BARTRUM has resigned the position of Chief Draughtsman of the plant construction department of Messrs. Dorman Long and Co., Ltd., Middlesbrough, Yorks, to take up the position of Chief Draughtsman at Messrs. John Miles and Partners (London) Ltd., Cannon Street, London, E.C.4.

▶ Mr. D. K. CHAKRAVORTY has been appointed Assistant Works Manager (under training), in the Indian Ordnance Service, at Kanpur, under the Ministry of Defence for the Government of India.

▶ Mr. K. C. CHOUDHURI has been elected an Associate of the Royal Institute of Chemistry.

▶ Mr. C. H. DAVY, a Director of Messrs. Babcock and Wilcox, Ltd., has been elected an Ordinary Member of Council of the Institute of Metals.

▶ Mr. R. DE BIE left the Carnegie Institute of Technology, of Pittsburgh, Pa., U.S.A., in June, to resume work with the Tréfileries Leon Bekaert, in Belgium.

▶ Mr. L. F. DENARO has returned from Zürich, and is now at The Mond Nickel Co., Ltd., Sunderland House, Curzon Street, London, W.1.

▶ Dr. C. H. DESCH, F.R.S., and Dr. R. DURRER, received the Honorary Degree of 'Dr. Mont.,' on the occasion of the celebration of the Centenary of the Montanistische Hochschule, in Leoben, Austria.

▶ Major H. B. DESHPANDE, of the metallurgical laboratory of the Locomotive Workshops (B.N. Railway), Kharagpur, India, has been elected a Corporate Member of the Chartered Institution of Engineers, India.

▶ Mr. E. L. DIAMOND resigned the position of Head of the Mechanical Section of the Plant Engineering Division of the British Iron and Steel Research Association, at the end of May, to take up the appointment of Assistant Technical Director, in charge of the Mechanical Industry Section, of the British Standards Institution.

▶ Mr. E. C. ELLWOOD, of the Department of Metallurgy, King's College, Newcastle-upon-Tyne, has been awarded the Degree of Doctor of Philosophy, by the University of London. He has also been awarded a travelling fellowship, by the Nuffield Foundation, to study extraction metallurgy in Rhodesia and South Africa.

▶ Mr. A. L. FEILD, of The Rustless Iron and Steel Corporation, Baltimore, Md., U.S.A., has been awarded the honorary degree of D.Sc. by the Stevens Institute of Technology, at Hoboken, New Jersey, U.S.A.

▶ Mr. F. B. FOLEY, Superintendent of Research of the Midvale Co., U.S.A., will deliver the Howe Memorial Lecture for 1950, at the American Institute of Mining and Metallurgical Engineers.

▶ Mr. G. GODDARD has taken up an appointment as senior scientific officer with the Ministry of Supply.

▶ Mr. E. J. HANCOCK has resigned the position of Deputy Blast-Furnace Superintendent of the Haripur and Kulti works of the Indian Iron and Steel Co., Ltd., to take up the position of Blast-Furnace Manager with the South Durham Steel and Iron Co., Ltd., at Hartlepool.

▶ Dr. V. J. D. HILL, Manager of Thermal Welding Products, Ltd., of South Africa, has been elected to represent South Africa, on the Brittle Fracture and on the Weldability Committees, of the newly formed International Institute of Welding.

▶ Mr. P. W. HYDE has left the research laboratories of the G.E.C., Ltd., Wembley, Middx., to join the Westlands Aircraft Co., Ltd., of Yeovil, Somerset.

▶ Mr. J. S. KERR, of the Lancashire Steel Corporation, Ltd., Irlam, near Manchester, has been awarded the Honorary Degree of M.Sc. of Manchester University.

▶ Mr. J. W. LAMBERT has recently been appointed Chief Chemist to the British Electro Metallurgical Co., Ltd., of Sheffield.

▶ Dr. F. C. LEA has been made an Honorary Member of the Institution of Mechanical Engineers.

▶ Mr. F. M. LOWE, a student at Cardiff University, has has been appointed Chemist-in-charge at the Royal Ordnance Factory, Nottingham.

▶ Sir ANDREW MCCANCE, F.R.S., has been awarded the E. J. Fox Medal, for 1949, by the Institute of British Foundrymen.

▶ Dr. A. B. MCINTOSH has been appointed Assistant Director, in the Division of Atomic Energy (Production), responsible for metallurgical development.

▶ Dr. A. G. QUARRELL, of the British Non-Ferrous Research Association, has been elected an Ordinary Member of Council of the Institute of Metals.

▶ Mr. H. SADLER, formerly billet mill manager at Messrs. S. Fox and Co., Ltd., of Stocksbridge, has been appointed Works Manager at the Templeborough Rolling Mills, Ltd., Rotherham.

▶ Mr. J. SANDERS is no longer Chief Metallurgist to the Vulcan Boiler and General Insurance Co., Ltd., Manchester, but has a consulting practice at Westfield Road, Wellingborough, Northants, under the name of J. S. Industrial Laboratories.

▶ Mr. F. SCOPES, Managing Director of the Stanton Ironworks Co., Ltd., Nottingham, has been appointed Chairman of the British Coking Industry Association, 11 Pall Mall, London, S.W.1.

▶ Mr. J. A. SHAW has been appointed Managing Director of Messrs. F. W. Southern and Co., Ltd., of Brighouse, Yorks; he is also Secretary of the Bradford Steel Pin Manufacturing Co., Ltd., of Bradford.

▶ Dr. C. J. SMITHELLS, Director of Research of the British Aluminium Company, Ltd., Gerrards Cross, Bucks, has been elected a Vice-President of the Institute of Metals.

▶ Mr. F. T. SISCO, of New York, Director of Alloys of Iron Research, was appointed, as from 1st March, 1949, Technical Director of the Engineering Foundation, the joint research agency of the American Institutes of Electrical Engineers, Mechanical Engineers, Chemical Engineers, and Mining and Metallurgical Engineers.

▶ Mr. F. M. STEARN has been appointed assistant metallurgist, in the Process Development Department of Messrs. Henry Wiggin and Co., Ltd., Birmingham.

▶ Mr. E. T. TURKDOGAN has been awarded the degree of Master of Metallurgy of Sheffield University.

▶ Mr. W. UDALL, of the Brightside Foundry and Engineering Co., Ltd., Sheffield, has joined the Board of Messrs. Buckley and Taylor Ltd., Castle Ironworks, Oldham, Lancs.

▶ Mr. W. B. WALLIS, of the Pittsburgh Lectromelt Furnace Corporation, of Pittsburgh, Pa., U.S.A., has been conferred Honorary Life Membership of the American Foundrymen's Association.

▶ Mr. A. H. WATERFIELD has left the Ministry of Supply

to take up an appointment as Scientific Attaché at the British Embassy in Paris, under the Department of Scientific and Industrial Research.

▶ Dr. J. H. WHITELEY has retired from the Consett Iron Co., Ltd., of Consett, Co. Durham.

▶ Professor H. K. WORNER has succeeded Professor J. N. GREENWOOD as Honorary Corresponding Member for Australia, to the Council of the Institute of Metals.

▶ Mr. C. A. ZAPFFE, Consulting Metallurgist, received honourable mention for his joint paper with Mr. M. E. HASLEM, on "Evaluation of Pickling Inhibitors from the Standpoint of Hydrogen Embrittlement," presented to the Wire Association of the U.S.A.

Obituary

Mr. W. J. PETERSEN, of Johannesburg, South Africa, on 6th October, 1948.

E. Sharratt

R. R. Roberts

CONTRIBUTORS TO THE JOURNAL

E. Sharratt, B.Sc., Ph.D., F.R.I.C.—Was senior research chemist at the Murex Welding Processes, Ltd., Waltham Cross. Dr. Sharratt, after graduating with first class honours in Chemistry in the University of Birmingham in 1933, carried out research on the "Stereochemistry of the Transition Elements" with Prof. W. Wardlaw, C.B.E., obtaining his Ph.D. in July 1936. He was then for four years on ceramics research at Stoke-on-Trent. In 1940, he was appointed to the research staff of Murex Welding Processes, Ltd., where he was senior research chemist at the time of his death. His publications include papers on platinum, palladium, and copper, on various ceramic subjects, and on size analysis.

It is very much regretted that Dr. Sharratt died on 11th March, 1949.

R. R. Roberts, M.A., L.I.M.—Metallurgist, Murex Welding Processes, Ltd., was educated at Peter Symonds School, Winchester, and Christ's College, Cambridge. He obtained the degree of B.A.(Cantab.) in 1945, and M.A. in 1949. From 1944 to 1946 he was experimental assistant, Central Metallurgical Laboratory, Admiralty. He took up his present position in 1947.

B. D. Burns, M.Sc.—Principal Scientific Officer at the Armaments Research Establishment at Woolwich. Mr. Burns graduated at University College of North Wales in Bangor, in 1936, with first class honours in Physics. He joined the Armaments Research Department at Woolwich in 1940, and subsequently transferred to the Metallurgical Department of the National Physical Laboratory, Teddington.

Mr. Burns took up his present position, as head of the Physics and X-ray Group of the Metallurgical Branch of the Armaments Research Establishment, in August 1948.

A. H. Cottrell, B.Sc., Ph.D. Lecturer in metallurgy at Birmingham University, and in charge of a research group, supported by the Department of Scientific and Industrial Research, working on problems of plastic deformation. Dr. Cottrell graduated at Birmingham University in 1939, and obtained the degree of Ph.D. in 1942. Since 1945 he has specialized in the field of physical metallurgy. He took up his present position in 1943.

A. T. Churchman, B.Sc.—Research graduate in theoretical physical metallurgy at Birmingham University. Mr. Churchman is working in a research group dealing with plasticity of metals and alloys, supported by the Department of Scientific and Industrial Research. Born in 1926 in Birmingham, and educated at King Edward's High School, he graduated with B.Sc., first class honours in theoretical metallurgy, at Birmingham University, in 1946.

E. T. Turkdogan, M.Met.—Research student at Sheffield University. Mr. Turkdogan was born in 1923, and educated at Izmir College, Turkey. In 1943, he won a State scholarship to Sheffield University, where, in 1947, he gained the degree of B.Met., first division. He has recently been awarded the degree of M.Met. by Sheffield University. In September, 1948, he received the Carnegie Research Fund Scholarship, for research work for the Ph.D. degree.

Audrey M. B. Douglas, M.A., Ph.D.—University Demonstrator in the Department of Mineralogy and Petrology, Cambridge. Dr. Douglas was educated at South Hampstead High School, and at Girton College, Cambridge, where she read for Parts I and II of the Natural Sciences Tripos. From July 1942, she worked in the Crystallographic Section of the Cavendish Laboratory, at Cambridge, on problems concerned with the iron–carbon system, notably, the structure of martensite, for which, together with other work on cobalt–aluminium, she obtained the degree of Ph.D., in 1948.

B. D. Burns

A. H. Cottrell

A. T. Churchman

E. T. Turkdogan

Dr. Douglas became Assistant Lecturer in Chemistry at Leeds University in October, 1945, and took up her present position in January, 1948.

R. W. Powell, D.Sc., Ph.D., Principal Scientific Officer, in the Heat Section of the Physics Division of the National Physical Laboratory. Dr. Powell was educated at Erith County School, and at University College, London. He graduated with B.Sc., first class honours in Physics, in 1924, and was awarded the degree of Ph.D., in 1938, and the degree of D.Sc., in 1944.

Dr. Powell is serving as Recorder of Section A (Mathematics and Physics), for the 1949 Meeting of the British Association for the Advancement of Science. He joined the Heat Section of the Physics Division of the National Physical Laboratory in 1924.

AFFILIATED LOCAL SOCIETIES

Swansea and District Metallurgical Society

In addition to the list of Officers published in the June issue of the *Journal* (*see* p. 227) the following members have been co-opted to the Council :

M. L. HUGHES, M.Sc., F.R.I.C., F.I.M.
H. PHILLIPS, M.I.Mech.E.

BRITISH IRON AND STEEL RESEARCH ASSOCIATION

The Physical Chemisty of Process Metallurgy

The proceedings of the meeting of the Faraday Society, held at Ashorne Hill from 23rd to 25th September, 1948, to discuss "*The Physical Chemistry of Process Metallurgy*," have now been published. In addition to the papers presented at the meeting, the complete discussions are reported. Copies may be obtained from the publishers, Messrs. Gurney and Jackson, London and Edinburgh, price 30*s*.

A brief report of the meeting, and a list of papers of particular interest to those concerned with iron and steel, was given in the February, 1949, issue of the *Journal* (*see* p. 142).

Sixth Refresher Course on Steel Making

The Steelmaking Division has arranged for the Sixth Refresher Course on Steel Making to be held in Glasgow from Monday, 4th July, to Friday, 8th July, 1949, inclusive.

The principal subject of discussion will be "Basic Refractory Materials used in Steel Making" with special reference to open-hearth practice. Considerable attention will be given to the qualities of the refractory materials demanded by steel manufacture, and the factors which control them. Slag attack and methods of combating this will be discussed. From a study of the constitutional relationships and recent work on the mechanism of solid solution formation, possible lines of future development will be considered. As a subsidiary subject of discussion, the diffusion of gases, directional factors in gas flow, and the influence of heated brick surface on heat transfer will be considered. Laboratory experiments will be set up to illustrate many of the topics under discussion.

Third Refresher Course on Iron Making

This course will be held at the Royal Technical College, Glasgow, under the supervision of Professor R. Hay, from 4th July to 8th July, 1949 (at the same time as the Steel Making Course).

The main subject to be studied will be "High Temperature Reactions in the Blast-Furnace with Particular Reference to Silicon and Sulphur," and the subsidiary subject will be "Principles of Operation of Hot-Blast Stoves." A works visit will be arranged in the afternoon of 6th July.

INSTITUTE OF METALS

Honorary Corresponding Members

The Council of the Institute of Metals (London) has appointed the following members to be Honorary Corresponding Members to the Council, for their respective countries :

Belgium : HENRI FÉRON, Administrateur-Directeur, Visseries et Tréfileries Réunis S.A., Haren.

Holland : MAX HAMBURGER, Director, Royal Nederlandsche Lood-en Zinkpletterijen voorheen A.D. Hamburger, Utrecht.

Italy : Dott. LENO MATTEOLI, Vice-Director, Instituto Scientifico Tecnico Ernesto Breda, Sesto S. Giovanni, Milano.

Capper Pass Awards

The Capper Pass Awards Adjudicating Committee has made the following awards for 1948, on behalf of the Councils of the Institution of Mining and Metallurgy and of the Institute of Metals.

C. BLAZEY, L. BROAD, W. S. GUMMER, and D. P. THOMPSON (Metal Manufacturers, Ltd., Port Kembla, N.S.W., Australia), £50 jointly, for a paper on "The Flow of Metal in Tube Extrusion," published in the *Journal of the Institute of Metals*, 1948, vol. 75, December.

H. R. POTTS (Minas de Rio Tinto), £50, for a paper on "Further Notes on Converter Practice at Rio Tinto," published in the *Bulletin of the Institution of Mining and Metallurgy*, 1948, March.

R. V. TRUMBELL (Pyrites Co., Inc.), W. HARDIEK (Pyrites Co., Inc.), and E. G. LAWFORD (Rio Tinto Co., Ltd.), £50 jointly, for a paper on "Notes on the Treatment of Pyrites Cinders at the Plant of the Pyrites Co., Inc., Wilmington, Delaware," published in the *Bulletin of the Institution of Mining and Metallurgy*, 1948, December.

The Capper Pass Awards, which are made annually from a sum of £200 per annum, placed at the disposal of the Councils of the Institution of Mining and Metallurgy and of the Institute of Metals by Messrs. Capper Pass and Sons, Ltd., Bristol, are as follows :

(*a*) £100 per annum, for one or more awards to authors of papers on some aspect of non-ferrous extraction metallurgy.

(*b*) £100 per annum, for one or more awards to authors of papers, relating to some process or plant used in the fabrication of non-ferrous metals, contributed by persons engaged full time in industrial practice.

The Adjudicating Committee will meet early in 1950, to consider all papers published by both societies during the year 1949. Papers on extraction metallurgy should be submitted, in duplicate, to the Secretary of the Institution of Mining and Metallurgy, Salisbury House,

Finsbury Circus, London, E.C.2, and papers on fabrication should be submitted to the Secretary of the Institute of Metals, 4 Grosvenor Gardens, London, S.W.1.

INSTITUTION OF METALLURGISTS
Metallurgical Exhibition

The Council has approved the holding of a Metallurgical Exhibition, sponsored by the Institution, to be held at the Science Museum, South Kensington, from July to September, 1950. The exhibits are to be arranged under the general title "*Metals in the Service of Mankind.*" Preliminary arrangements are now in hand.

It is intended that the Exhibition shall include a modern well-equipped metallurgical laboratory, and a historical survey of discovery, extraction, and processing of all the metals down to modern times. In addition, the most modern methods of production and refining, and the most modern applications of metals will be dealt with in sections allotted to each metal or group of metals.

Members of the Institution are asked to interest themselves in this project and should feel free to offer suggestions, which will be welcomed and considered by the Working Committee.

NEWS OF SCIENCE AND INDUSTRY

Fourth Empire Mining and Metallurgical Congress

The Third Circular (April 1949) has now been distributed. It gives the final programmes in the various centres, a list of papers to be read, and a list of the delegates who propose to attend the Congress, which is being held in Great Britain from 9th–23rd July, 1949.

Members who are interested, and who have not already made application, should apply for further information to the Joint General Secretaries, Fourth Empire Mining and Metallurgical Congress, 436, Salisbury House, Finsbury Circus, London, E.C.2.

Cambridge Summer School in Electron Microscopy, 1949

By the courtesy of Professor Sir Lawrence Bragg, O.B.E., F.R.S., a Summer School in Electron Microscopy will be held again this year in the Cavendish Laboratory, Cambridge.

The school will be held from 16th to 27th August, 1949, inclusive. A detailed syllabus may be obtained from G. F. Hickson, M.A., Secretary of the Board of Extra-Mural Studies, Stuart House, Cambridge.

Engineering and Marine Exhibition

The next of this series of exhibitions, formerly known as the Shipping, Engineering, and Machinery Exhibition, and revived in 1947, is to be held at Olympia, London, W.14, from 25th August to 10th September, 1949.

The Organizers have extended, to all members of the Institute, an invitation to visit the Exhibition, and requests for complimentary tickets should be made to the Secretary, 4 Grosvenor Gardens, as soon as possible.

United Nations Conference on the Conservation and Utilization of Resources

The United Nations Scientific Conference on The Conservation and Utilization of Resources, will be held at Lake Success, U.S.A., from 17th August to 6th September, 1949. The aim of the conference is to take the first step in a world mobilization of the resource techniques. Particulars of the conference may be obtained from the Joint Secretaries of the C.S.R. Committee, Mr. H. L. Verry and Mr. G. A. McMillan, D.S.I.R., Africa House (3rd floor), Kingsway, London, W.C.2.

Congrès Technique International de l'Entreprise de Peinture en Bâtiment et des Industries Annexes

The first International Technical Congress of the structural painting industry and of allied industries, will be held in Paris from 19th to 22nd October, 1949, under the patronage of the Minister of Reconstruction and Town Planning, Paris. The Congress will include an exhibition of supplies for the profession. The exhibition will be open from the 17th to 22nd October, 1949.

Second Oil Shale and Cannel Coal Conference

The Institute of Petroleum have announced the proposal to convene the Second Conference on Oil Shale and Cannel Coal, to meet in Glasgow during the week commencing 3rd July, 1950. The opportunity is being taken to celebrate, by a Young Memorial Lecture, the centenary of the James Young patent which led to the foundation of the Scottish shale-oil industry.

The technical sessions of the Conference will be in three sections: (*a*) Geology and Mining, (*b*) Retorting, Refining, and Uses of By-Products, and (*c*) Economics and Statistics.

Further details may be had on application to the Secretary of the Institute of Petroleum, 26 Portland Place, London, W.1.

Industrial Finishes Exhibition, 1950

This Exhibition will be held at Earls Court, London, S.W., in September, 1950, and not in August to September, 1949, as announced in the *Journal* for March, 1949, p. 251. After consultation, it was decided that the time which had been allowed for the preparation of the Exhibition was not sufficient, either for the exhibitors, or for the Advisory Technical Committee, charged with the preparation of a Central Technical Exhibit designed to provide up-to-the-minute information for manufacturers.

It is proposed to hold, concurrently, a Scientific Congress on Industrial Finishing, at which the foremost experts on the subject will speak. It is also proposed to issue a Handbook, which will serve as an up-to-date guide for manufacturers, to the best finish available from the range of finishes produced by British scientific skill.

University Scholarship Schemes, 1949

The British Coking Industry Association are offering up to ten scholarships for University courses suitable for careers in the coking industry. The awards are available to young men already employed in, or connected with, the industry, and to school and university students. All candidates must be British subjects, and those from schools will be expected to pass, or to have passed, the Higher School Certificate Examination, or its Scottish equivalent, with a high standard in mathematics, physics, and chemistry. Those from universities must be taking suitable courses and be recommended by their professors. Those from the industry should be under twenty-five years of age and have reached university entrance standard.

The Scholarships are tenable at the Universities in London, Birmingham, Cardiff, Durham, Glasgow, Sheffield, Leeds, Manchester, and Nottingham.

Particulars may be obtained from the Secretary, the British Coking Industry Association, 11–12 Pall Mall, London, S.W.1.

Scholarship—University of Durham

The Power-Gas Corporation, Ltd., of Stockton-on-Tees, has endowed an Open Scholarship, tenable at Durham University, to enable suitably qualified students to undertake a course of study, leading to a degree, in either civil or mechanical engineering. The Scholarship will normally be tenable for a period of three years but, in exceptional cases, it may be extended for a fourth year.

In 1947, the Power-Gas Corporation endowed a similar Open Scholarship at the University of Leeds, for a course in gas, chemical, mechanical, or civil engineering.

Research in Physical Sciences

The Pressed Steel Co., Ltd., of Cowley, Oxfordshire, have offered the sum of £10,000 a year for a period of seven years, to Oxford University, for the encouragement of research in physical sciences.

British Oxygen Company's Prize for Welding Research

No award of the British Oxygen Company's Prize for Welding Research has been made for 1948/1949, but it is intended to offer the Prize again this year, when it is hoped that some welding research, deemed worthy, will be forthcoming.

Candidates should submit an original unpublished paper on a research into welding or its application, not exceeding 5000 words, carried out in Great Britain or Northern Ireland.

Papers submitted for the competition should reach the Secretary, British Welding Research Association, 29 Park Crescent, London, W.1, on or before 30th September, 1949.

New Research Laboratories at General Refractories Ltd.

The new research laboratories of General Refractories Ltd., at Sandy Lane, Worksop, Notts., were opened on 26th April, 1949, by Sir Charles Goodeve, F.R.S., Director of the British Iron and Steel Research Association. A plaque, erected in the entrance hall, to the memory of the first chairman of the firm, Mr. Frank S. Russell, was unveiled by the present chairman, Sir Ronald Matthews.

National Foundry College

Notice of the new session of the National Foundry College, at Wolverhampton, which commenced in September, 1948, was given in the December issue of the *Journal*, pp. 431–432.

In April 1949, a hostel was opened for students of the Course. The staff of the College has been increased. Two full-time lecturers, in addition to the Head of the College, have been appointed, and the appointment of a third lecturer is contemplated.

Of the 18 students in the first Diploma session, 16 gained the Diploma, 4 of them with distinction. Of these students, 15 had been connected with iron founding, 2 with steel founding, and 5 with non-ferrous metal founding. The number of students in the second session, beginning in September 1948, is 18, of whom 12 have had their main experience in iron founding, 4 in steel founding, and 5 in non-ferrous metal founding.

The next Diploma Course, Part 2, will open in September, 1949, for ten months' duration, and the preliminary course, Part I, will open about the same date, to be of six months' duration, thus permitting several intervening months of foundry practice for students who propose to follow Part I by the succeeding Part 2.

Memoranda

Changes of Address

▶ The Coal Tar Research Association moved to their new Research Station, at Oxford Road, Gomersal, near Leeds, on Monday, 30th May, 1949.

▶ The head office of Messrs. H. J. Enthoven, of Upper Thames Street, London, is now at Forum House, Lime Street, London, E.C.3. The departments formerly at Harrow, Croydon, and Shortlands, are also at the new address as from 2nd May, 1949.

DIARY

4th–8th July—BRITISH IRON AND STEEL RESEARCH ASSOCIATION—Third Refresher Course on Iron Making—"High Temperature Reactions in the Blast-Furnace, with Particular Reference to Silicon and Sulphur," with "Principles of Operation of Hot-Blast Stoves," as subsidiary subject—under the supervision of Professor R. Hay—Royal Technical College, Glasgow.

6th July—THE INSTITUTION OF MINING AND METALLURGY—Sir Julius Wernher Memorial Lecture. "*The Effect of Impurities on the Properties of Metals*," by Dr. C. H. Desch, F.R.S.—The Royal Institution, 21 Albemarle Street, London, W.1, 5 P.M.

7th–8th July—INSTITUTION OF MINING AND METALLURGY—Symposium on the Refining of Non-Ferrous Metals—The Royal Institution of Chartered Surveyors, Great George Street, London, S.W.1, 10 A.M. to 5 P.M.

9th–23rd July—FOURTH EMPIRE MINING AND METALLURGICAL CONGRESS—Great Britain.

10th–15th July—FOURTH WORLD POWER CONFERENCE—The Institution of Civil Engineers, Great George Street, London, S.W.1.

13th July—INSTITUTE OF WELDING—Annual General Meeting—The Institution of Civil Engineers, Great George Street, London, S.W.1, 2.30 P.M.

17th Aug.–6th Sept.—UNITED NATIONS SCIENTIFIC CONFERENCE on "The Conservation and Utilization of Resources"—Lake Success, U.S.A.

TRANSLATION SERVICE

(The previous announcement was made in the June, 1949, issue of the *Journal*, p. 228.)

TRANSLATIONS AVAILABLE

No. 380 (German). K. GUTHMANN: "Carburetting in Open-Hearth Furnaces by Means of Gas." (*Stahl und Eisen*, 1949, vol. 69, Mar. 31, pp. 223–227).

TRANSLATION IN COURSE OF PREPARATION

(Russian). M. I. KOROBOVA and N. I. KOROBOV: "Catalytic Agents for the Reduction Process in Blast-Furnaces." (*Bulletin de l'Académie des Sciences, URSS, Classe des Sciences Techniques*, 1946, No. 4, pp. 567–577).

CHARGES FOR COPIES OF TRANSLATIONS—For the above translations a charge will be made of £1 for the first copy and 10*s*. for each additional copy of the same translation. Requests should be accompanied by a remittance. The above translations are not available on loan from the Joint Library.

TRANSLATIONS PREPARED AT MEMBERS' REQUEST—Members requiring translations of foreign papers are invited to communicate with the Secretary, who will ascertain whether the translations can be prepared for inclusion in the Series.

ABSTRACTS OF CURRENT LITERATURE *and* BOOK NOTICES

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MINERAL RESOURCES

Geology of the Northern Pennine Orefield.—Tyne to Stainmore. K. C. Dunham. (Department of Scientific and Industrial Research, Memoirs of the Geological Survey of Great Britain, 1948, vol. 1 : H.M. Stationery Office).

The Differentiation Problem of the Apatite Iron Ores. B. Asklund. (Geologiska Föreningens i Stockholm Förhandlingar, 1949, vol. 71, No. 1, pp. 127–176). [In Swedish]. After giving a comprehensive review of the geology and petrology of the apatite-bearing iron ores, especially those of the Kiruna type, the author establishes that they are differentiation products of a lower order from special parental rock magmas and eruptive provinces with always recurring distinct chemical characteristics.—R. A. R.

The Vähäjoki Iron Ore in Tervola, Northern Finland. A. Mikkola. (Bulletin de la Commission Géologique de Finlande, 1947, No. 140, pp. 261–280 : [Abstract] Centre National de la Recherche Scientifique, Bulletin Analytique, 1948, vol. 9, No. 10, p. 2459).

Development of the Iron Ores in the Aalenian and Bajocian Strata of the Middle Rhine Valley (Alsace, Baden). N. Théobald. (Mem. Serv. Carte Géol. Alsace Lorraine, 1948, No. 8 : [Abstract] Centre National de la Recherche Scientifique, Bulletin Analytique, 1948, vol. 9, No. 10, p. 2459).

Poland's Coal Deposits. G. V. Schenk. (Báňský Obzor, 1948, vol. 2, July, pp. 105–111). [In Czech]. The geographical distribution of the individual coal and lignite deposits is described and qualitative data on the various deposits are given. The output of the Polish coal mines was about 58 million tons in 1913 ; during the war (1939–45) the yearly output was about 80 million tons, but it dropped to about 21 million tons in 1945. It increased to 59 million tons in 1947 and it is planned to mine 77·5 million tons in 1949. E. G.

Iron Ores Reserves in Michigan. F. G. Pardee. (Mining and Metallurgy, 1948, vol. 29, Nov., pp. 613–614). In a recent survey the iron ore reserves of Michigan are estimated at 392,000,000 long tons (direct-shipping ore and those concentrates produced by methods now in commercial use or likely to be used in the near future). This figure is made up as follows : Gogebic range 112,000,000 long tons, Menominee range 140,000,000 long tons, Marquette range 140,000,000 long tons. Some notes are given on the location of these deposits.—J. C. R.

ORES—MINING AND TREATMENT

Some Ore Prospecting Instruments. D. Malmqvist. (Instruments and Measurements Conference, Stockholm, 1947, Transactions, pp. 242–249). A brief account is given of three relatively new prospecting instruments which have been fully tested in practice in the north of Sweden by the Boliden Mining Co. They are a gravimeter for determining differences in gravity, equipment for prospecting for electromagnetic ore by the 'loop-frame' method, and an inductive angle indicator for measuring the dip of diamond-drill holes. R. A. R.

The Use of Anthracite Coal in Sintering Iron Ore at the Bethlehem Steel Co. Concentration Plant, Lebanon, Pa. A. R. Walter. (Transactions of the Sixth Annual Anthracite Conference, Lehigh University (U.S.A.), 1948, pp. 139–158 ; [Abstract] Monthly Bulletin of the British Coal Utilisation Research Association, 1948, vol. 12, Dec., p. 452). Tests were made with anthracite and with coke and anthracite mixtures. Too coarse coal leads to waste of fuel ; too fine coal does not mix well with the ores. The ash contents must be low.

Preparation of Mesabi Ore for a 2000 Ton Per Day Blast Furnace. S. Naismith. (Blast Furnace and Coke Association of the Chicago District and Eastern States Blast Furnace and Coke Oven Association, Joint Meeting : Blast Furnace and Steel Plant, 1948, vol. 36, Dec., pp. 1476–1479). After discussing trials with screened and agglomerated Mesabi ores, and noting improved production and reduced flux and coke consumption, the author considers the possibility of preparing the ores in a similar fashion for a 2000-ton/day blast-furnace with a hearth diameter of 28 ft., and concludes that the success in the trials was largely due to the high grade ore employed. He points out that (1) the chemical analysis of future Mesabi ores makes it doubtful if they could be improved to the grade required ; (2) the trend of shipments shows a decreasing iron-to-silica ratio, whilst improvement in structure is due to the shipping of increasing quantities of concentrates ; and (3) the beneficiation would involve the agglomeration of 20,000,000 tons/annum of fines below 20 mesh and require large capital expenditure.—J. P. S.

FUEL—PREPARATION, PROPERTIES, AND USES

Development of a Possible Method of Estimating the Excess Air Used in Combustion of Blast-Furnace and Coke-Oven Gas from a Knowledge of Gross Calorific Value of the Fuel and the Carbon Dioxide Content of the Waste Gases. H. J. Victory. (Coke Oven Managers' Association : Gas World, 1949, vol. 130, Apr. 2, pp. s-7–s-24).

Combustion Control. H. Ziebolz. (Yearly Proceedings of the Association of Iron and Steel Engineers, 1948, pp. 747–751). *See* Journ. I. and S.I., 1949, vol. 161, Apr., p. 372.

Controls for Multiple Fuels. W. F. Pray. (Yearly Proceedings of the Association of Iron and Steel Engineers, 1948, pp. 579–583). *See* Journ. I. and S.I., 1949, vol. 161, Jan., p. 56.

Use of Oxygen-Enriched Air Jets Around an Oil Burner. M. W. Thring. (Iron and Coal Trades Review, 1949, vol. 158, Mar. 11, pp. 509–510). Assuming that the amount of pure oxygen which can be used is fixed by economic considerations, the author attempts to predict the optimum amount of air which should be added around an oil jet for furnace heating. He estimates that the use of enriched air containing between 22% and 60% of oxygen is likely to be as beneficial or more beneficial than enriched air containing over 60% of oxygen.—J. C. R.

The Third Symposium on Combustion Flame and Explosion Phenomena. G.Whittingham. (Monthly Bulletin of the British Coal Utilisation Research Association, 1948, vol. 12, Dec., pp. 429–438). The author reviews the papers presented at the Third Symposium on Combustion, Flame and Explosion Phenomena, held at the University of Wisconsin on 7–11 September, 1948. For the purpose of this review the papers are divided under seven main headings: (1) The Ignition of Gas Mixtures; (2) Flame Propagation and Stabilization; (3) Kinetics and Spectroscopy of Flames and Combustion; (4) Combustion in Engines and Rockets; (5) Burning of Explosives; (6) Thermodynamics and Thermochemistry of Flame Gases; and (7) Experimental Techniques.—R. F. F.

Modern Design of Multiple Fuel Steam Units. A. R. Mumford. (Yearly Proceedings of the Association of Iron and Steel Engineers, 1948, pp. 156–166). *See* Journ. I. and S.I., 1948, vol. 158, June, p. 210.

Design and Performance of a Refractory Recuperator. E. G. Smith. (Yearly Proceedings of the Association of Iron and Steel Engineers, 1948, pp. 295–300). *See* Journ. I. and S.I., 1948, vol. 159, Aug., p. 434.

Heat Transfer. A. D. Davidson. (Society of Chemical Industry, Chemical Engineering Group, Nov., 1948, Advance Copy). This paper approaches the subject from the viewpoint of the designer of industrial heat transfer equipment, and examines the origin, nature, and significance of some of his problems. It draws attention to the limitations of the present design tools, discusses their improvement, and suggests new ones.—R. A. R.

Theory of Coal Sampling. A. B. Manning. (Fuel, 1949, vol. 28, Mar., pp. 49–56). The accuracy with which a sample of coal represents the bulk can be calculated from the weight of the sample taken, the particle-size distribution, and the results of a float and sink test. Conversely, the minimum weight of sample required for a given accuracy can be calculated. The theory is developed from the calculated distribution in samples drawn at random from a mixture of two different kinds of particles, *e.g.*, particles of coal and particles of dirt, and thence extended to apply to ternary mixtures. The essential formulæ are given. Some typical examples are worked out.

Sampling of Coal and Washery Products. J. Visman. (Transactions of the Fuel Economy Conference, The Hague, 1947, vol. 2, pp. 430–441). For any sort of coal and for any desired accuracy, the weight of sample and the number of increments can be calculated if the 'sampling constants' of that coal are known. The use of experimental data concerning the weight of sample and the number of increments in connection with the ash content and particle size are no longer needed when applying the specification described in this paper.—R. A. R.

A Technical Study of Coal Drying. G. A. Vissac. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 2537: Mining Engineering, 1949, vol. 1, Feb., Section 3, pp. 56–60). Various aspects of coal drying are dealt with, mainly from a mathematical viewpoint.—J. C. R.

Recent Developments in Coal Washing. M. G. Driessen. (Transactions of the Fuel Economy Conference, The Hague, 1947, vol. 2, pp. 462–481). Modern processes of coal preparation, with special reference to heavy-media separation, are described. Whereas exact separation in the low-gravity ranges is technically possible with larger coal above ¼ in., methods of cleaning fine coal by heavy media are still in course of development. A description is given of the new cyclone washer operating by heavy media on fine coal but using centrifugal force instead of gravity. An appendix contains general information on the theory of cyclones and tornadoes.—R. A. R.

The Blending of Coals to Improve Coke and Extend Coking Resources. P. S. Savage. (Blast Furnace and Steel Plant, 1949, vol. 37, Mar., pp. 323–324, 334). Diminishing reserves of high quality coking coals can be extended first by careful blending, at the coke ovens, of high and low sulphur coals, and secondly by improvements in the desulphurization of blast-furnace iron and the removal of sulphur from coke-oven gas. Practice is described in the blending of anthracite fines and of weathered strip coals with strongly coking coals, which has given cokes of quality equal to those made from unblended coals.—J. P. S.

The Use of Illinois Coal in the Production of Metallurgical Coke. F. H. Reed, H. W. Jackman, and P. W. Henline. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 2491: Coal Technology, 1948, vol. 3, Nov.). An account is given of an investigation of the coking properties of blends of low-sulphur, high-volatile Illinois coal with high- and low-volatile coals from the eastern states of America. The coke was produced in an experimental, small scale, slot-type oven (capacity, 500 lb. of coal) under conditions whereby the coke would have similar physical and chemical properties to commercially produced coke. Data derived from experimental results are given.—J. C. R.

Coke-Oven Plant. H. C. Wood. (British Engineering, 1949, vol. 31, Feb., pp. 1298–1303). Illustrations and descriptions are given of the latest designs of British-made mobile or travelling machinery for serving coke-oven batteries. R. A. R.

National Tube Builds New Coke Plant at Lorain. (Yearly Proceedings, of the Association of Iron and Steel Engineers, 1948, pp. 301–303). *See* Journ. I. and S.I., 1948, vol. 159, July, p. 322.

The Modern Coke Plant. T. J. Ess. (Yearly Proceedings of the Association of Iron and Steel Engineers, 1948, pp. 54–88). *See* Journ. I. and S.I., 1948, vol. 159, May, p. 84.

Dust Collection. (Iron and Steel, 1949, vol. 22, Mar., pp. 98, 100). A description is given of the Musgrave Aerodynamic Dust Collector (Système Linderoth) the filter surface of which consists of a specially slotted sheet bent into the form of a cone. The air or gas to be cleaned is passed over this surface at high velocity and, due to the form of the slots, aerodynamic forces are set up, the resultant force of the dust particles being in a direction away from the cone face. This force keeps the dust particles suspended in the form of a thin cloud layer immediately in front of the filter surface while the clean air escapes through the slots. The dust layer is rapidly carried to the outlet end of the cone where it is continuously withdrawn, together with a very small percentage of air or gas, into a secondary circuit. The concentrated dust mixture in the secondary circuit is then passed through a small collector into a suitable container, the secondary air or gas being returned to the inlet of the cone where it again passes through the same process.—J. C. R.

Present State of Fuel Oil in the Steel Industry. (Petroleum Times, 1947, vol. 51, Dec. 20, pp. 1262–1264: [Abstract] Centre National de la Recherche Scientifique, Bulletin Analytique, 1948, vol. 9, No. 11, p. 3139). This is an economic study of the advantages of using liquid fuels in the iron and steel industry.

REFRACTORY MATERIALS

Shales, Clays and Kaolins in France.—I. M. Larcheveque. (Industrie Céramique, 1948, No. 385, p. 75: British Ceramic Abstracts, 1948, Nov., p. 339A). This first part of a comprehensive review deals with the refractory clays and shales. These are divided into seven classes, four according to locality (Fumel, Provins, Langeais, Vandœuvre-en-Brenne) and three according to constitution (micaceous, ferruginous and micaceous or calcareous, and kaolinitic). Methods of examining clays are described.

Ohio Fire Clays as Foundry Bond Clays. N. J. Dunbeck. (Ohio State University, Engineering Experiment Station News, 1947, vol. 19, No. 2, p. 32: British Ceramic Abstracts, 1948, Nov., pp. 339A–340A). Fireclay of the Lawrence vein in southern Ohio is equal to the Missouri fireclays in refractoriness and is superior in bonding value; it is superior in both respects to the fireclays of Illinois, Indiana, West Virginia, Pennsylvania, and North and East Ohio. Ball clays of Kentucky, Tennessee, and New Jersey have high wet strength but inadequate baked strength. The greater strength of the Lawrence clay reduces the quantity used,

lessens the water required and gives both higher permeability and flowability. The durability of this clay is higher than most fireclays. The bulk of it is obtained by strip mining, that procured from underground being harder and weaker. Whilst bonding clays occur abundantly in Lawrence County, lesser quantities are found in Jackson and Scioto Counties. Owing to transportation difficulties development will take many years. The bond clay combined with flint clay is also used in producing firebricks. Bentonites are widely used in foundries on account of their special properties. The southern bentonites have three times the green strength of fireclays, but lower dry strength, whereas the western bentonite has double the green strength and higher dry strength. Each is selected for its ability to do a particular job, although sometimes they are used in combination.

Refractories from Ohio Dolomite. H. C. Lee. (Ohio State University, Engineering Experiment Station News, 1947, vol. 19, No. 2, p. 38 : British Ceramic Abstracts, 1948, Nov., p. 342A).

Haydite Aggregate for Refractory Purposes. W. C. Smith. (Journal of the Canadian Ceramic Society, 1948, vol. 17, p. 69 : British Ceramic Abstracts, 1948, Nov. p. 360A). A brief review is given of the general characteristics of "Haydite" as a refractory material. It is used for semi-refractory and insulation purposes up to 540° C. and, in conjunction with ciment fondu, up to 1200° C. Fusion point is 1310° C., above which the strength falls off rapidly. Sodium silicate improves the strength at around 820° C. Some qualities and characteristics of Haydite alumina concrete are high insulation, high strength, light weight, minimum spalling, adaptability as poured concrete or as blocks, etc., low expansion and shrinkage, and slag resistance, whilst prefiring is unnecessary. Its uses include furnace-door linings, arches, crowns, etc.

Raw Materials for the Manufacture of Silica Bricks for Coke-Ovens, Etc. L. Hachet. (Industrie Céramique, 1948, No. 385, p. 80 : British Ceramic Abstracts, 1948, Nov., pp. 359A–360A). The author considers that the specifications imposed by the French coke-oven industry are too stringent ; these demand a specific gravity not greater than 2·34, a silica content not less than 93–94% and a refractoriness-under-load of not less than 1650° C. If a silica content of 90% were allowed, and a refractoriness-under-load of 1550°–1600° C., the French reserves of suitable silica rock would be greatly increased. It is argued that such bricks would still be adequate to meet service conditions in coke ovens, for bricks of this quality are giving satisfaction in the glass industry.

Refractory and Heat-Resistant Concrete. S. B. MacDonald. (Journal of the Canadian Ceramic Society, 1948, vol. 17, p. 72 : British Ceramic Abstracts, 1948, Nov., p. 360A). A brief outline is given of the principle involved in the use of refractory concrete. Consideration is given to factors affecting strength, spalling, and the ability to manufacture special shapes or use as poured-in material. Heat-resistant concrete is used under conditions of exposure to continued soaking, such as in the foundations of blast-furnaces. Insulating concrete, with a conductivity one-third that of firebrick, is used for service up to 540° C., but above this refractory grogs are incorporated.

Solubility Relationships of the Refractory Monocarbides. J. T. Norton and A. L. Mowry. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 2527 : Journal of Metals, 1949, vol. 1, Feb., Section 3, pp. 133–136). The purpose of this investigation was to examine the solubility of the several pairs of the carbides of titanium, zirconium, vanadium, niobium, and tantalum, by heating them together until equilibrium was established, and then examining the product by X-rays. Experimental results showed clearly the limit placed on solubility by the size factor. The binary carbide systems TiC–ZrC, TiC–VC, TiC–TaC, ZrC–NbC, TiC–NbC, ZrC–TaC, VC–NbC, VC–TaC, and NbC–Ta C form a continuous series of solid solutions. The binary carbide system ZrC–VC shows very small solubility, the limits at 2100° C. being 5% at the zirconium-rich end and less than 1% at the vanadium-rich end. The lattice-parameter/composition curves show only slight deviations from Vegard's law, but both positive and negative deviations are found. The maximum size difference between atom diameters expressed in per cent of the solvent atom diameter for extended solubility is 13%. The minimum value for very restricted solubility is 16%. In the solid solution the metal atoms are distributed at random on the points of the lattice.

New Heat-Resistant Materials that Slash Process Costs. (Modern Ind., 1948, vol. 16, No. 4, pp. 40–45 : American Ceramic Abstracts, 1949, vol. 32, Jan. 1, p. 15) Pure oxides, carbides, nitrides, borides, and metal oxides are special refractories for higher furnace temperatures and rocket temperatures approaching 4000° F. A table gives melting points, top operating temperatures in oxidizing and reducing atmospheres, relative resistance to heat, shock, weight, approximate cost, use, and remarks regarding slag resistance, conductivity, resistivity, etc., for super-duty fireclay, super-duty silica, electrically fused mullite, chromite, forsterite, fused alumina, silicon carbide, beryllia, zircon, zirconia, fused magnesia, thoria, and carbon.

Rapid Method of Determining Specific Gravity of Quartzites and Dinas Products. V. D. Tsigler. (Ogneupory, 1948, vol. 13, No. 3, pp. 131–133 : American Ceramic Abstracts, 1949, Jan. 1, p. 21).

Common Types of High Frequency Melting Furnaces. A. R. Baffrey. (Charleroi-Jeumont, 1947, Nos. 1/2, p. 1 : British Ceramic Abstracts, 1948, Nov., p. 363A.) After outlining the theory of the high-frequency induction furnace, the construction of such units is described with brief notes on the types of refractory lining employed. It is recommended that basic linings should be rammed dry at a temperature above 100° C.

Destruction of Silica Furnace Chamber Brick in Wet Service by Salt-Containing Steam. H. Bruckner. (Gas- und Wasserfach, 1947, vol. 88, No. 3, p. 89 : American Ceramic Abstracts, 1949, Jan. 1, p. 14).

Refractory Materials in the Foundry Industry. G. R. Rigby and A. T. Green. (Proceedings of the Institute of British Foundrymen, 1947–48, vol. 51, pp. A144–A151). *See* Journ. I. and S.I., 1948, vol. 160, Dec., p. 436.

BLAST-FURNACE PRACTICE AND PRODUCTION OF PIG IRON

Investigations Relating to Scaffolds in Blast Furnaces. G. R. Rigby. (Journal of The Iron and Steel Institute, 1949, vol. 161, Apr., pp. 295–300). Investigations have been made on a number of scaffolds occurring in blast-furnaces burdened with home ores. These scaffolds presented certain similarities and, as a result of laboratory examinations, a general theory for the formation of scaffolds has been postulated. A scaffold found in one furnace exhibited certain marked differences, and suggestions have been made as to the variations in blast-furnace practice which may have resulted in producing these modifications.

The Inactive Central Core in the Blast-Furnace. E. Terlecki. (Hutnik, 1948, vol. 15, May–June, pp. 202–204). [In Polish]. The causes of the inactive central core in the blast-furnace are given as low blast pressure, tuyeres of too large diameter, and insufficient tuyeres. An analysis of a sample of a green powder taken from this core in the bosh of a dismantled furnace had the following result : Zn 31·46%, S 0·47%, Pb 1·37%, CaO 13·88%, MgO 8·79%, SiO_2 6·30%, Fe 6·24%, balance 31·49%.

Progress in High Pressure Operations of Blast Furnaces. B. S. Old, E. L. Pepper and E. R. Poor. (Yearly Proceedings of the Association of Iron and Steel Engineers, 1948, pp. 305–315). *See* Journ. I. and S.I., 1948, vol. 159, Aug., p. 434.

Coke Oven Expansion and Blast Furnace Operation in 1948. C. Longenecker. (Blast Furnace and Steel Plant, 1949, vol. 37, Jan., pp. 67–70). Advances in blast-furnace practice in 1948 have included the conversion of a number of furnaces to high top pressure operation, the installation of two blast-furnaces of a capacity of 1500 tons a day each, and consideration of the construction of a furnace for a capacity of 2000 tons a day. Because of the shortage of native ores, iron ores are being imported from Canada, Sweden, and South America. On the coking side it is considered that the supply of good coking coal is rapidly diminishing, and beneficiation must be extended. The construction of coke ovens is expanding, however, and 1124 ovens for which contracts were let in 1948 were operating or nearly completed by early January 1949.—J. P. S.

PRODUCTION OF STEEL

Power and Power Factor in Arc Furnace Operation. E. H. Browning. (Yearly Proceedings of the Association of Iron

and Steel Engineers, 1948, pp. 13–20). *See* Journ. I. and S.I., 1948, vol. 159, May, p. 87.

Killed Bessemer—A New Steel of High Quality. E. G. Price. (American Iron and Steel Institute, Sept. 29, 1948: Metal Progress, 1949, vol. 55, Jan., pp. 39–42). The author describes the process of manufacture of deoxidized acid-Bessemer steel. In this process molten iron is added at the end of the blow so that the carbon in the iron removes the oxygen from the steel as a gas. Ferromanganese and ferro-silicon are added, and finally 3 lb. of aluminium are added per ton of steel in the ladle to ensure complete deoxidation. This steel is suitable for the manufacture of seamless pipe and for oil-well casing, where resistance to collapse from external pressure is desired. The properties of this steel at elevated temperatures are similar to aluminium-deoxidized open-hearth steel, while the yield strength is usually 10,000–15,000 lb./sq. in. higher.

Hot-Pressed 8·8 and 12·8 cm. Anti-Aircraft High Explosive Made from Basic-Bessemer Steel with 0·25–0·35% C. H. Wübbenhorst and H. Hauttmann. (Verein deutscher Eisenhüttenleute, 1944, Confidential Report No. 79). Comprehensive manufacturing, firing, and bursting tests were carried out on 8·8 and 12·8-cm. anti-aircraft high-explosive shell with low carbon content, yield point and tensile strength. The carbon was 0·25–0·35% compared with the usual 0·55–0·70%, the yield point 30 kg./sq. mm. against 40, and the tensile strength 50 kg./sq. mm. against 75. The advantages in melting are lower spiegel consumption (1 kg./ton), and greater recovery from the scrap. The consumption of cutting tools was less in the machining operations, and output was increased owing to permissible greater speed of feed. The new steel is better from the point of view of fragmentation, for the number of fragments heavier than 30 g. is five to ten greater. No greater deformation of the shell on firing was observed. This could be expected from the known influence of velocity on resistance to deformation. The whole production of anti-aircraft high-explosive 8·8 and 12·8 cm. shells was turned over to the new steel. Brief tests with basic-Bessemer rimming steel of the same strength showed promise.—G. W. A.

Multiple Fuel Burners for Open Hearth Furnace. J. M. Brashear. (Yearly Proceedings of the Association of Iron and Steel Engineers, 1948, pp. 328–336). *See* Journ. I. and S.I., 1948, vol. 160, Oct., p. 220.

Recent Trends in Open-Hearth Furnace Design. A. H. Leckie. (Sheffield Metallurgical Association, Feb. 11, 1949: Iron and Coal Trades Review, 1949, vol. 158, Mar. 4, pp. 465–467). Recent developments in open-hearth furnace design are briefly discussed, mainly with reference to port design.—J. C. R.

Developments in the Design of Open-Hearth Furnaces. J. Alexandrovský. (Convention of Mining Engineers and Metallurgists, 1947: Báňský Obzor, 1948, vol. 2, Aug., p. 127). [Abstract in Czech]. Reducing the fuel consumption per ton of steel by increasing the size of furnace is discussed and the developments of open-hearth furnaces in the U.S.A., the U.S.S.R., Great Britain, Canada, and Germany are compared. The trend is towards building furnaces of increasing size, and in the U.S.S.R. 13·4% of their furnaces are units of over 150 tons. The Guthmann diagram is used to show the relation between the increase in output and the increase in furnace size. Use of cold charges reduces the output. The output of an open-hearth furnace depends on the fuel used and is higher with coke-oven gas than with producer gas; good results were obtained in Germany during the war with "*Restgas*" obtained in the production of synthetic petrol. Comparison of the specific fuel consumption of a 50-ton furnace and two 25-ton furnaces show that the saving in fuel alone is enough to pay off the cost of one furnace in ten years. Further improvements are to be sought by finding the optimum ratio between the hearth surface and the furnace capacity. The influence of this ratio on the initial investment and the operational costs is represented by graphs. The author emphasizes that, especially in this field, improvements can be made on many of the existing installations. The design of furnace parts is dealt with in detail and some guiding principles to be followed by designers to obtain high efficiency in open-hearth furnaces are enumerated.—E. G.

The Modern Open Hearth. T. J. Ess. (Yearly Proceedings of the Association of Iron and Steel Engineers, 1948, pp. 479–530). *See* Journ. I. and S.I., 1948, vol. 160, Oct., p. 220.

Weirton Steel Company. T. J. Ess. (Yearly Proceedings of the Association of Iron and Steel Engineers, 1948, pp. 362–380). *See* Journ. I. and S.I., 1948, vol. 160, Sept., p. 103.

The Layout of Integrated Iron and Steel Works. H. H. Mardon and J. S. Terrington. (Journal of The Iron and Steel Institute, 1949, vol. 161, Apr., pp. 327–359). The nature and extent of the problem of the layout of an integrated iron and steel works is considered and factors are discussed concerning the selection of a site, the layout of an integrated works as a whole, and some of the criteria which determine the detailed arrangements of each of the three departments of ironmaking, steelmaking, and rolling mills. Features existing in works in Great Britain are tabulated; a theoretical analysis of a fundamental theory of the layout of integrated iron and steel works is developed; and a comparative study of the traffic routes of the main process materials in the principal integrated works in Great Britain is made possible by flow diagrams reproduced in colour. Owing to the complexity of the details, it is not possible in many instances to lay down precise rules to suit all conditions, but some general conclusions have been drawn. Numerous aspects require detailed quantitative analysis and the most important of these is considered to be a traffic survey of representative works.

Basic Ends and All-Basic Open Hearth Furnaces. R. P. Heuer and M. A. Fay. (Yearly Proceedings of the Association of Iron and Steel Engineers, 1948, pp. 181–194). *See* Journ. I. and S.I., 1948, vol. 159, June, p. 215.

Modern Melting Shop. (British Engineering, 1949, vol. 31, Feb., pp. 1329–1331). This is a description of the melting shop of the Appleby-Frodingham Steel Co., Ltd. (*See* Journ. I. and S.I., 1949, vol. 161, Apr., p. 373).—R. A. R.

The Conversion to Oil Firing of the Open-Hearth Furnaces at Park Gate Works. D. F. Marshall and H. C. White. (Journal of The Iron and Steel Institute, 1949, vol. 161, Apr., pp. 301–317). The paper gives an account of the experience gained during the conversion of a complete open-hearth furnace plant to the use of fuel oil. After briefly describing the open-hearth plant the reasons for making the conversion are discussed and some consideration is given to the properties of the principal available fuels. The details of the fuel-oil plant and the methods adopted in making the conversion are outlined. A comprehensive system of instrumentation has been used to apply a simplified operating technique to hand-controlled furnaces. The methods of furnace control adopted and the organization required to ensure that the best use is made both of furnaces and of instruments, are given in some detail. The operating results over a period of 57 weeks are presented and discussed under three main headings, shop layout, type of charge, and furnace performance. Under the last heading consideration has been given to the quality of the scrap and to the constitution of the charge. The flow characteristics of three types of burner and details of the influence of roof temperature on the rate of furnace working are given. The life of the refractories used is indicated; and the essentials for fully automatic furnace control are stated.

Oxygen. A Review of Its Applications in the Iron and Steel Industry. W. C. Newell. (Iron and Steel, 1949, vol. 22, Mar., pp. 77–80). After discussing generally the use of oxygen in the iron and steel industry the author reviews the application of its addition to blast-furnaces, cupolas, open-hearth furnaces, electric-arc furnaces, and Bessemer converters. A bibliography of 103 references is appended.—J. C. R.

Oxygen in the Steel Industry. H. W. McQuaid and F. E. Pavlis. (Yearly Proceedings of the Association of Iron and Steel Engineers, 1948, pp. 237–244). *See* Journ. I. and S.I., 1948, vol. 159, July, p. 325.

Oxygen and Steel. D. Murray. (Scope, 1949, Mar., pp. 48–57, 104). The author draws a picture of the iron and steel industry as it may develop as a result of the extensive use of oxygen or oxygen-enriched air. It is suggested there would be no need for blast-furnace stoves, the blast-furnace could be reduced in height and consequently lower grade fuels with less resistance to crushing could be used. In open-hearth practice a lower quality fuel like blast-furnace gas could be used with oxygen, and regenerators could be reduced in size and eventually done away with.—R. A. R.

The Use of Oxygen in Open Hearth Furnaces. E. C. McDonald. (Yearly Proceedings of the Association of Iron and Steel Engineers, 1948, pp. 171–177). *See* Journ. I. and S.I., 1948, vol. 159, June, p. 215.

Operation of Statistical Quality Control in a Steel Mill. W. T. Rogers. (Steel, 1949, vol. 124, Mar. 14, pp. 102, 105, 108, 111, 114–122). The applications of correlation analysis in a steelplant including blast-furnaces, open-hearth and Bessemer plant, rolling mills, and pipe mills are discussed. Examples are given which include investigations of the effect of variables on the weight per foot of seamless pipe, on the efficiency of manganese in blocking practice in the open-hearth furnace, and the effect of carbon and manganese on the hardenability index of a steel.—J. P. S.

Developments in Open Hearth Practice. C. R. FonDersmith. (Blast Furnace and Steel Plant, 1949, vol. 37, Jan., pp. 71–73). In 1948, more than 81,000,000 ingot tons were produced by the open-hearth process; the chief interests of operators have been in securing sufficient scrap, in investigating alternative fuel supplies in face of erratic sulphur contents of fuel oil, in the use of hot blown metal in the furnace, and in basic construction. The use of oxygen is now considered to be of the greatest help in carbon removal; in high hot metal charges it gives little help as an aid to combustion.—J. P. S.

Production of Open-Hearth Steel. L. F. Reinartz. (Steel, 1949, vol. 124, Jan. 17, pp. 76–84, 87, 90, 92, 95; Jan. 24, pp. 68–72, 75–80; Jan 31, pp. 82–91: Feb. 7, pp. 106–118; Feb. 14, pp. 96–106). The first part of this article deals with the history of the open-hearth process in the United States, and discusses the design and the refractories of modern basic furnaces, using oil fuel or mixed oil and gas. In the second part the arrangements for stocking and charging iron ore, limestone, and scrap are described, together with hot-metal ladles and charging procedure. The working of a charge in the furnace is treated in the third part and pouring practice considered for both killed and rimming steels. In the fourth part, the chemistry of the basic and acid processes is explained, and the production and application of both killed and rimming steels is treated from this point of view. The final part deals with the combustion system in an oil-fired "cross regenerative" furnace, and describes the controls and instruments required for combustion efficiency.—J. P. S.

The Manufacture of Chromium Steels in the Basic Open Hearth and Electric Arc Furnaces. F. Badenheuer, W. Heischkeil, and H. Müller. (Verein deutscher Eisenhüttenleute, 1943, Confidential Report No. 51). Experience with various processes for the production of chromium steels from charges high in chromium under conditions favourable for output and the separation of phosphorus and chromium with, if necessary, removal of sulphur from the pig iron by soda between blast-furnace and open-hearth plant, are described. The usual basic electric arc and duplex processes are classified and compared, with data on the charges, chromium recovery, and output.—G. W. A.

The Manufacture of Tungsten Steels Using Alloy Scrap and Ore. F. Badenheuer. (Verein deutscher Eisenhüttenleute, 1941, Confidential Report No. 28). The manufacture of tungsten steels in the basic open-hearth, employing only alloy scrap containing tungsten or tungsten ores, by a specially developed procedure is described. This permits the efficient utilization of tungsten previously attained only in the electric furnace. The basic open-hearth furnace is not suitable from the point of view of metal economy for the production of high-tungsten steels, in contrast to the acid open-hearth, which enables the production of high tungsten steels up to, say, 20% of tungsten, to be carried out. The acid furnace is particularly adapted to working up hydrated and rusty scrap, as well as alloying from tungsten ore. A process which renders possible the treatment of increased quantities of chromium or chromium scrap is also described. The investigations on the replacement of ferrotungsten by manganiferous tungsten ores in basic and acid electric furnace have shown that the manufacture of tungsten steels in the basic furnace is not completely satisfactory from the point of view of metal economy. On the other hand, the manufacture of tungsten steels from alloy scrap and manganiferous tungsten ores alone in the acid high-frequency furnace is very favourable from metal recovery and general economic standpoints.—G. W. A.

The Manufacture of Chromium Steels Using Chromium-Bearing Scrap in the Basic Open-Hearth and the Electric Furnace. F. Badenheuer. (Verein deutscher Eisenhüttenleute, 1942, Confidential Report No. 39). Critical analysis of a large number of chromium-bearing melts made in the basic open-hearth enabled the conditions which influence the retention of chromium to be determined. By the application of special processes developed in the Krupp steelworks, it is possible to retain 0·9% to 1·4% of chromium in the steel with a mean charge content of 1·3% to 1·9%, which is several times the figure given in the literature, *i.e.*, 0·4%. The necessary conditions are detailed, especially those of minimum slag volume and high temperature. Various working processes in the basic arc furnace, distinguished by the level of the chromium in the charge and the chromium yield, are described with special reference to the quality of the steel produced. Recoveries of 75–95% of the chromium in the scrap are possible. The influence of the phosphorus content on the chromium yield if an oxidation period is interposed is dealt with. Figures are given on chromium losses in the slags and the mean chromium recovery from alloy scrap at a large steelworks producing chromium steels where one half the chromium content was provided by scrap.—G. W. A.

Application of the Molten Slag Process in Duplexing. H. Stallmann. (Verein deutscher Eisenhüttenleute, 1944, Confidential Report No. 54). Tests on rapid dephosphorization by the Perrin process were carried out with a slag made from lime, ore, and fluorspar. An oil-fired (later altered to grid gas-fired) rotating-drum furnace lined with stamped dolomite was used to melt the slag. One ton of slag, with 50–60% of lime and 15–20% of iron, with a height of drop of 2·5–0·6 m. was used for 23 tons. The removal of phosphorus, nitrogen, and sulphur is described. After a brief comparison of the time saved and increased output of the Duplex process as against solid charges, operation with basic oxidizing slags is described. Since the lining problem was not satisfactorily solved, comparative Duplex charges were made with solid slag constituents charged into the Girod furnace. The advantages were almost the same as in the Perrin process. The process should therefore be used by works operating the Duplex process.—G. W. A.

The Duplex Process. D. T. Rogers. (Yearly Proceedings of the Association of Iron and Steel Engineers, 1948, pp. 245–254). *See* Journ. I. and S.I., 1948, vol. 159, July, p. 325.

Basic Arc Furnace Processes in the Light of the 'Semi-Duplex' Process. R. Rinesch. (Verein deutscher Eisenhüttenleute, 1943, Confidential Report No. 41). The well-known and most used arc-furnace processes are described, and the new semi-duplex process is compared with them from the points of view of recovery of the alloy elements, rate of output, electrode and fuel consumption, and steel quality. The process consists essentially of pouring molten metal from a primary open-hearth furnace on a heated solid charge of heavy and light alloy scrap (20–40% of total). The charge is melted and finished under reducing conditions. There is no boil. The plant described consisted of two 20-ton open-hearths, 4 tons/hr. each, and a 20-ton arc furnace (charged to 24 tons), output 190 tons/24 hr. with a consumption of 350–400 kWh./ton.—G. W. A.

Slab and Billet Conditioning Symposium—Pneumatic Hand-Chipping. P. S. Paluch. (Yearly Proceedings of the Association of Iron and Steel Engineers, 1948, pp. 430–431). *See* Journ. I. and S.I., 1948, vol. 160, Oct., p. 221).

Slab and Billet Conditioning Symposium—Application of Hand Scarfing to Modern Surface Conditioning. P. S. Paluch. (Yearly Proceedings of the Association of Iron and Steel Engineers, 1948, pp. 431–435). *See* Journ. I. and S. I., 1948, vol. 160, Oct., p. 221.

Slab and Billet Conditioning Symposium—Billet Conditioning with Mechanical Chippers. J. H. Vollmer. (Yearly Proceedings of the Association of Iron and Steel Engineers, 1948, pp. 435–441). *See* Journ. I. and S.I., 1948, vol. 160, Oct., p. 221.

FOUNDRY PRACTICE

1938–1948, Ten Years of Progress in British Ironfounding. W. W. Braidwood. (Fonderie, 1948, Dec., pp. 1401–1416: Foundry Trade Journal, 1949, vol. 86, Mar. 24, pp. 253–260; Mar. 31, pp. 283–288). A review is presented of trends in the British ironfounding industry over the last ten years.—J. C. R.

The Addition of Oxygen to the Cupola Blast. J. Blakiston. (Institute of British Foundrymen: Foundry Trade Journal. 1949, vol. 86, Mar. 31, pp. 289–290). Some of the advantages and disadvantages of adding oxygen to the cupola blast are discussed. Reference is made to the importance of the cupola stack height and the modification of cupola design.—J. C. R.

Role of Calcium Silicide in the Refining of Iron. C. Bordoni. (Annali di Chimica Applicata, 1947, vol. 37, Oct.–Nov., pp. 412–416: [Abstract] Centre National de la Recherche

Scientifique, Bulletin Analytique, 1948, vol. 9, No. 10, p. 2376). The mechanism of the action of calcium silicide on liquid iron is discussed. The best conditions were temperature > 1350° C. and a silicon content of 0·3%.

Melting Stainless Steel in an Induction Furnace. G. L. Meeter. (Foundry, 1948, vol. 76, Dec., pp. 97, 249). The author describes experiments to determine what alloy additions should be made to titanium stabilized stainless steel scrap of known composition which, when melted in an induction furnace, produce stainless steel castings of specified composition.—J. C. R.

The Mechanics of Moulding Sand Properties. A. Jamieson. (Foundry Trade Journal, 1949, vol. 86, Mar. 3, pp. 183–184). The author examines the physical properties of sands with a view to finding a means of measuring their mouldability. It is suggested that the ratio of plastic flowability to granular flowability would offer a good indication of the mouldability of a sand.—J. C. R.

Patterns and Molding Methods for Steel Castings. J. H. Hall. (Foundry, 1948, vol. 76, Nov., pp. 80–83, 212–226; Dec., pp. 92–95, 209–216; 1949, vol. 77, Jan., pp. 76–79, 208, 210, 213; Feb., pp. 90–93, 210–222; Mar., pp. 92–97, 260–262). The characteristics of various types of pattern equipment are outlined and a description is given of the conditions under which each type is to be preferred and the moulding methods to which each is best adapted.—J. C. R.

Synthetic Resin Core Binders. G. L. Harbach. (Institute of British Foundrymen: Foundry Trade Journal, 1949, vol. 86, Mar. 10, pp. 203–208; Mar. 17, pp. 225–233). A comprehensive review is given of the use of synthetic-resin core binders as a substitute for oil binders, and the properties of resin-bonded cores. Among the factors considered are baking time and temperature, rate of gas evolution during the pouring and solidification of the casting, moisture absorption, knock-out properties, and surface finish of the casting.—J. C. R.

Core Baking—Two Minutes *versus* Four Hours. V. E. Hillman. (Iron Age, 1949, vol. 163, Feb. 3, pp. 116–121). A dielectric core-baking unit has been installed by the Crompton and Knowles Loom Works, Worcester, Mass., for baking small cores; the thermosetting binder is also simultaneously polymerized. A 30-kW., 15-megacycle electronic oscillator is used. The cores remain in the heating zone 2–2½ min. and reach a temperature of 275° F. (135° C.); 400–600 lb. of cores can be dried per hour.—J. P. S.

Handling Materials in Metal Charging. R. H. Herrmann. (Foundry, 1948, vol. 76, Dec., pp. 80–85, 218, 221, 224). Modern equipment used for charging metals in foundry melting departments is described and illustrated.—J. C. R.

Mechanization of Westinghouse Foundry. A. D. Stout, jun. (Iron Age, 1949, vol. 163, Jan. 13, pp. 54–57: Foundry Trade Journal, 1949, vol. 86, Mar. 3, pp. 185–187). An illustrated description is given of the grey iron foundry of Westinghouse Electric Co. at East Springfield, Mass., which has been recently redesigned, and produces motor-housing castings up to 40 lb. on jolt-squeeze roll-over-type machines and small castings of 3 lb. which are moulded on small automatic machines.—J. C. R.

A Reconstructed Foundry. V. C. Faulkner. (Foundry Trade Journal, 1949, vol. 86, Mar. 17, pp. 235–237). A description is given of the plant and operations at the Butlin Foundry, Wellingborough, where precision grey-iron castings are produced.—J. C. R.

HEAT-TREATMENT AND HEAT-TREATMENT FURNACES

The Heat Treatment of Iron and Steel: Recalescence. L. F. Keeley. (Machinery Lloyd, 1948, vol. 20, July 3, pp. 68–72: [Abstract] Bulletin and Foundry Abstracts of the British Cast Iron Research Association, 1949, vol. 9, Jan., p. 351). This is a survey of the fundamental mechanisms of heat-treatment and the accompanying phase transformations.

Heating and Treatment of Alloy Steels. A. H. Arbogast and M. K. Morris. (Iron and Steel Engineer, 1949, vol. 26, Jan., pp. 75–81). An illustrated description is given of five bell-type gas-fired furnaces installed at the Bethlehem works of the Bethlehem Steel Co. for the heat-treatment of high-speed and air-hardening grades of tool steels. The furnaces were designed for direct firing with desulphurized coke-oven gas, with the combustion system equipped with a two-zone temperature programme control. The function of the inner bell is to protect the charge from scaling and decarburization; this is effected by circulating a mixture of nitrogen and propane under the bell during the annealing cycle. The bell is fitted with an oil seal to prevent infiltration of outside air or the products of combustion into the treatment atmosphere. Data are given of operational results with the furnaces.—J. C. R.

New Methods for Surface Hardening Gears. F. Pobořil. (Báňský Obzor, 1948, vol. 2, July, pp. 97–104; Aug., pp. 116–125; Sept., pp. 136–141; Oct., pp. 152–158). [In Czech]. The author has developed a simple, cheap and effective method for surface-hardening gears of carbon and alloy steels of nearly eutectoid composition. According to this method, called the "AC 1 method," the gears are heated in thermal baths to temperatures of 10° to max. 100° C. above the Ac_1 point of the particular steel. The heating time is chosen so that the gears remain in the heating bath just long enough to allow the temperature of the teeth surfaces to rise above the Ac_1 point, whilst the temperatures of the teeth cores and other massive parts of the gears remain below that point. The heated gears are then quenched in thermal baths at about 200° C., and a hard surface layer is obtained, whilst the cores remain soft. The toughness of the cores can be considerably increased by heat-treatment before applying the Ac_1 surface hardening. Compared with flame-hardening, the author's method is simpler and more effective; fully automatic controls can be used to control bath temperatures and heating times. Compared with high-frequency methods, this process does not require expensive apparatus, and gears of various sizes and of different steels can be handled without change in the equipment. As the temperatures are not excessively high, wear on the pots is small. Internal stresses are small and cracking very infrequent because the difference in temperature between surface and core is small. Gears hardened by the new method were tested in the gear box of a modern lathe and showed no appreciable wear after 4608 hr. service.—E. G.

Economics of Radio-Frequency Heating. A. P. Bock. (Steel, 1949, vol. 124, Feb. 21, pp. 94–97: Industrial Heating, 1949, vol. 16, Jan., pp. 230–238, 323, 324). The factors contributing to the cost of radio-frequency heating are discussed and graphs are presented showing the relation between production cost and rate of production, using equipment of various powers.—J. P. S.

Heating for Hardening and Forging with R.F. Equipment. T. E. Lloyd. (Iron Age, 1949, vol. 163, Feb. 17, pp. 86–92). International Harvester Co., Louisville, Kentucky, uses high-frequency heating for hardening gears and crankshafts. A frequency of 60 cycles is used for preheating heavy gears and one of 9600 cycles for heating to hardening temperature; lighter gears are preheated with 9600 cycles and heated for hardening with 250,000 cycles. Special gears and cylindrical surfaces are heated with 9600-cycle power and spray-quenched with water; small miscellaneous parts are heated with 450,000-cycle power and water-quenched. Induction heating for forging is done with 3000 cycle power; the process is used in upsetting and forging shafts, gears, crankshafts, and connecting rods.—J. P. S.

Versatility of Induction Heating. J. D. Graham, H. F. Kincaid and R. E. McGee. (Steel, 1949, vol. 124, Jan. 31, pp. 62–64, 80). The use made of induction heating at various frequencies by the International Harvester Co., at Louisville, Kentucky, is described (*see* preceding abstract).—J. P. S.

Designing for High-Frequency Induction Hardening. J. T. Temin. (Transactions of the American Society of Mechanical Engineers, 1949, vol. 71, Feb., pp. 135–146). The metallurgical aspects of high-frequency induction heating of steel parts are discussed with particular reference to the period of heating and the effect of frequency and increasing the energy input on the hardness depth. Consideration is then given to stationary, progressive, continuous, and oscillatory heating methods and the kinds of parts for which each method is best suited. Illustrations are given of various induction heating machines that have been designed for use with these methods.—J. C. R.

An Analysis of Nitriding. H. E. Boyer. (Iron Age, 1949, vol. 163, Feb. 10, pp. 68–74: Feb. 17, pp. 93–98). The first part of this article discusses the composition of nitriding steels, the rôle of aluminium in providing the hardening constituent, the function of other elements such as chromium and molybdenum in providing a tough core, and the necessity for heat-treating steels before nitriding. The second part

deals with the nitriding of austenitic steels and discusses the effect produced on corrosion resistance by the consumption of chromium in the formation of nitrides; the danger of carbide precipitation at the nitriding temperature is also indicated. The author concludes with a general discussion on equipment for nitriding and on the best ways of economically operating the process.—J. P. S.

Roughen Stainless before Nitriding. S. Dedomineco. (Machinist, 1949, vol. 92, Feb. 26, pp. 1422–1423). Tests on American stainless steels AISI 410 and 416 show that steels which have a roughened surface, such as produced by sand-blasting, pickling, or heat-treatment in a bright annealing furnace with an Ammogas atmosphere, will have a uniform and adequate case after nitriding. Stainless steels, the surface of which has been ground or machined, cannot be nitrided satisfactorily.—R. F. F.

Surface Hardened Stainless Steels. V. T. Malcolm. (Product Engineering, 1949, vol. 20, Jan., pp. 84–87). A nitriding process to harden the surface of stainless steel is described and the properties of the hardened surface are discussed. In this process, which is called 'Malcomizing,' dissociated ammonium is used to reduce any oxide on the surface of the steel, after which ionized ammonia is applied to add nitrogen to the surface. The steel is given a heat-treatment (usually by heating slowly to about 1700° F., soaking at that temperature, and quenching in water) before Malcomizing. Case depths up to 0·020 in. and a surface hardness up to Rockwell 94 can be obtained.—R. A. R.

Some Gas Applications in Steel Mills. H. H. Feirabend. (Industrial Gas, 1948, vol. 26, May, pp. 8–9: Fuel Abstracts, 1948, vol. 4, Nov., p. 199). This is a description of some of the newer steel mill applications, since the war, including such operations as continuous bright gas normalizing of strip steel, continuous bright gas annealing of strip steel, one-way-fired soaking pits, atmosphere annealing covers, gas carbon restoration, gas quenching, gas pickling, and continuous high temperature (2500 deg.) roller hearth furnaces.

Heat Transmission in Strip-Coil Annealing. J. D. Keller. (Yearly Proceedings of the Association of Iron and Steel Engineers, 1948, pp. 784–794: Iron and Steel Engineer, 1948, vol. 25, Nov., pp. 60–70). The author discusses heat transmission problems connected with the annealing of coiled strip and considers that practice has been determined by the convenience of handling the coils without regard to the best method of heating. He disputes the theory that metallic contact or radiation is responsible for heat transmission, and suggests that heat transfer is derived by conduction through the gas film and that this is the most important factor. More rapid heating and cooling might be effected if a gas with high thermal conductivity, such as hydrogen or helium, could be used. Other suggested solutions are the use of a circulating fan, heating by electrical resistors within hollow spaces between the coils, and passing the coils continuously, with their axes horizontal, through a tunnel furnace with a protective gas atmosphere.—J. C. R.

Hardening Capacity of Engineered Metal Castings. C. R. Austin. (Steel, 1949, vol. 124, Feb. 21, pp. 102–106). The hardening of Meehanite by quenching is discussed, and the factors affecting the critical cooling velocity are shown to be: (1) Time and temperature of holding above the critical range, and (2) composition. On the critical cooling velocity depends the depth to which the material will harden; the function of Jominy end-quench tests in establishing this is shown. The types of Meehanite metal which should be heat-treated are listed.—J. P. S.

The Hardening Operation in the Practical Heat Treatment of Steel. Q. C. McMillan. (West of Scotland Iron and Steel Institute, Feb., 1949, Preprint). The heat-treatment plant for dealing with large quantities of alloy steel bars at the Clyde Alloy Steel Co., Ltd., is described and data are presented on the physical properties of chromium–molybdenum and nickel–chromium–molybdenum steel bars obtained after various forms of heat-treatment.—R. A. R.

American Views on the Quenchability Properties of Steel. Y. Dardel. (Centre de Documentation Sidérurgique, Circulaire d'Informations Techniques, 1948, vol. 5, No. 6, Oct. 25, pp. 231–406). The whole of this issue is devoted to a survey of American views and concepts on the quenchability of steels. It is in two parts, each divided into a number of chapters. The first part deals with quenchability (the kinetics of the decomposition of austenite) under the headings: (1) General principles of phase changes; (2) isothermal decomposition of austenite in the pure-iron/carbon system; (3) the influence of alloying elements on the isothermal decomposition of austenite; (4) diagram of isothermal transformations; (5) anisothermal decomposition of austenite; and (6) diagram of anisothermal transformations. The second part is concerned with the measurement of quenchability: (1) Measurement of quenchability by means of round specimens; (2) Measurement of quenchability by means of Jominy specimens; (3) determination of the quenchability on the basis of the composition; (4) measurement of the susceptibility to tempering; and (5) determination of the domain of the formation of martensite on the basis of the composition. Six appendices deal with the general laws of the cooling of solid bodies, a simplified determination of the critical diameter, Anglo-American and French standards for the Jominy test, T.T.T. diagrams, and the limitations of the use of Grossmann's methods. The bibliography contains 164 entries.

A. E. C.

FORGING, STAMPING, DRAWING, AND PRESSING

The Production of Forgings for Marine Purposes. R. C. Benson. (North-East Coast Institution of Engineers and Shipbuilders, Feb., 1949, Advance Copy). The author outlines differences between castings, forgings, and rolled products, and points out the advantages claimed for forgings. The limitations encountered in producing forgings are explained and illustrated descriptions are given of a variety of marine forgings made by the Darlington Forge Co. Ltd.

R. A. R.

Theory of Flow Stresses in Metals at Drawing. S. I. Goobkin. (Bulletin de l'Académie des Sciences, U.R.S.S., Classe des Sciences Techniques, 1947, No. 12, pp. 1663–1681 [in Russian]: [Abstract] Applied Mechanics Reviews, 1948, vol. 1, Aug., p. 219). The author analyses critically several existing formulæ expressing the flow stress k in terms of the ultimate strength, the angle α characterizing the drawing, the coefficient of external friction μ, and the reduction per draw. He regards the formula of Körber (1940) as the best one, but points out that it does not take into account all the components of the resistance of drawing, and can therefore be applied to drawing solid bars, but not tubes. The author derives new formulæ for the stress from the assumption that, in a first approximation, this stress is a sum of three components k_1, k_2, and k_3, which, in accordance with simplified technological methods, are defined in a different manner from that used in the mathematical theory of plasticity. Also a new formula for the reduction per draw and a new determination of external friction are given. The figures are burdened by conspicuous errors, and obscure. The formulæ deduced are compared with the author's and other experimental results, and improved by means of empirical coefficients when necessary.

The Manufacture of Wrought Steel Wheels. W. A. Ashton and R. N. Merk. (Yearly Proceedings of the Iron and Steel Engineer, 1948, pp. 907–918). *See* Journ. I. and S.I., 1949, vol. 162, May, p. 116.

Modern Cold Draw Equipment. G. W. Garwig and A. L. Thurman. (Yearly Proceedings of the Association of Iron and Steel Engineers, 1948, pp. 770–780). *See* Journ. I. and S.I., 1949, vol. 161, Apr., p. 379.

ROLLING-MILL PRACTICE

Practices and Trends in Steel-Mill Control. A. W. Schmitz. (General Electric Review, 1948, vol. 51, May, pp. 34–37: Electrical Engineering Abstracts, 1949, vol. 52, Feb., p. 53). Mill equipment can be grouped into (1) auxiliaries, (2) processing lines, and (3) main roll drives. (1) Auxiliaries consist of independent single-motor drives, normally with magnetic time delay acceleration. Amplidyne control is used for hot-mill tables, sideguards, and screw-down drives. (2) Processing lines are driven by adjustable voltage-controlled D.C. motors. Automatic regulation of storage loops, strip tension, motor speed, etc., is frequently employed. (3) Amplidyne control is mainly used for the main roll drives of reducing mills, and typical schemes are illustrated; electronic control is not generally used owing to danger of shut-down by tube failure. A new amplidyne control with superimposed electronic amplifier vernier control is described, which can operate without the electronic portion and overcomes the shut-down danger.

J & L Rolls Tinplate at 70 Miles per Hour. (Yearly Proceedings of the Association of Iron and Steel Engineers, 1948, pp. 167–170). An illustrated description is given of the pickling plant and 42-in. five-stand tandem cold strip mill at the Aliquippa works of the Jones and Laughlin Steel Corporation.—R. A. R.

Operation and Maintenance of Automatic Preset Plate Mill Screwdown Control—Electrical Features. J. F. Skalka. (Yearly Proceedings of the Association of Iron and Steel Engineers, 1948, pp. 447–450). *See* Journ. I. and S.I., 1948, vol. 160, Oct., p. 226.

Operation and Maintenance of Automatic Preset Plate Mill Screwdown Control—Mechanical Features. R. G. Uhler. (Yearly Proceedings of the Association of Iron and Steel Engineers, 1948, pp. 450–452). *See* Journ. I. and S.I., 1948, vol. 160, Oct., p. 226.

Republic's New 48-in. Mill at Youngstown, Ohio. T. J. Ess. (Yearly Proceedings of the Association of Iron and Steel Engineers, 1948, pp. 671–673). *See* Journ. I. and S.I., 1948, vol. 161, Mar., p. 261.

Republic's 44–36–32 in. Mills at Chicago. R. F. Lavette. (Yearly Proceedings of the Association of Iron and Steel Engineers, 1948, pp. 717–721). *See* Journ. I. and S.I., 1949, vol. 162, May, p. 117).

Crucible Completes $3,200,000 Plant Improvement Program. (Steel, 1949, vol. 164, Feb. 28, pp. 90, 93). A new cold rolling plant has been erected for the Crucible Steel Co. of America, at Harrison, N.J.; it includes 15 new cold rolling mills to roll down to 0·004 in. thick at 1200 ft./min. Heat-treatment facilities comprise lead baths and gas furnaces; pickling arrangements cover both batch and continuous pickling. A special products division will manufacture permanent magnets, alloy steel tools, and other articles by precision casting with the 'lost wax' process.—J. P. S.

The Y-Mill—A New Type of Cold Strip Milling. A. B. Montgomery and W. M. McConnell. (Yearly Proceedings of the Association of Iron and Steel Engineers, 1948, pp. 229–234). *See* Journ. I. and S.I., 1948, vol. 158, June, p. 219.

Factors in the Operation of a Mannesmann Piercing Mill. H. R. McLaren. (Yearly Proceedings of the Association of Iron and Steel Engineers, 1948, pp. 261–269). *See* Journ. I. and S.I., 1948, vol. 160, Oct., p. 224.

MACHINERY FOR IRON AND STEEL PLANT

The Electrical Equipment of the Blast Furnace and Steel Works at Ijmuiden. T. Michielsen. (Ingenieur, 1948, vol. 60, Nov. 12, pp. 117–125 [in Dutch]: Electrical Engineering Abstracts, 1949, vol. 52, Mar. p. 102). Three blast-furnaces are in continuous operation, and supply five open-hearth furnaces of the steelworks. The rolling-mill plant is designed to produce plate of $\frac{1}{8}$–1 in. thickness. Separate power supply at 50 kV. from the local power station to each works is stepped down to 6 kV. and distributed by double busbar switchboards to the works' substations, where it is stepped down to 380 V. or fed direct to motors of ratings > 120–160 h.p. Cranes and transport machinery in the blast-furnace and steelworks are run on 550-V. D.C. from three rotary converters and one 100-kW. traction-type mercury-arc rectifier. The mill plant comprises a roughing mill which has twin drive by two 1000-h.p. moters at 25 r.p.m. and a 4-high finishing mill, driven by one 5500-h.p. motor at 65 r.p.m. The mill motors are fed through underground solid copper conductions of 12 kA. capacity from a Ward-Leonard-Ilgner set. Handling machinery in the mill is driven by 230-V. D.C. mill-type motors. The rollers of the roll-tables are direct-driven by 4-pole 110-V. D.C. shunt motors.

Dynamic Braking Magnetic Hoist Controller. K. S. Kuka. (Yearly Proceedings of the Association of Iron and Steel Engineers, 1948, pp. 889–894). *See* Journ. I. and S.I., 1949, vol. 162, May, p. 117.

Bearing Construction and Performance Characteristics. E. Crankshaw and K. Scheucher. (Yearly Proceedings of the Association of Iron and Steel Engineers, 1948, pp. 687–697). *See* Journ. I. and S.I., 1949, vol. 161, Mar., p. 262.

Steel Plant Service Requirements for Cranes. I. E. Madsen. (Yearly Proceedings of the Association of Iron and Steel Engineers, 1948, pp. 674–677). *See* Journ. I. and S.I., 1949, vol. 161, Mar., p. 261.

Operation of Ingot Buggies by Remote Control. K. S. Kuka. (Yearly Proceedings of the Association of Iron and Steel Engineers, 1948, pp. 421–423). *See* Journ. I. and S.I., 1948, vol. 160, Oct., p. 226.

Trends in Design and Application of Lifting Magnets. J. D. Leitch. (Yearly Proceedings of the Association of Iron and Steel Engineers, 1948, pp. 346–359). *See* Journ. I. and S.I., 1948, vol. 160, Sept., p. 109.

Crane Block, Limit Switch and Hoist Cable Failures. F. W. Cramer. (Yearly Proceedings of the Association of Iron and Steel Engineers, 1948, pp. 895–901). *See* Journ. I. and S.I., 1949, vol. 162, May, p. 117.

Circular Saws. R. D. Brooks. (Mechanical Engineering, 1949, vol. 71, Feb., pp. 133–138). Problems in the design, manufacture and operation of standard saws for all-purpose work are discussed.—J. C. R.

LUBRICANTS AND LUBRICATION

Lubricants for Cold Rolling Steel. R. J. Nekervis and R. M. Evans. (Yearly Proceedings of the Association of Iron and Steel Engineers, 1948, pp. 752–761). *See* Journ. I. and S.I., 1949, vol. 161, Apr., p. 380.

Water Lubrication of Phenolic Bearings. F. Vogt. (Blast Furnace and Steel Plant, 1949, vol. 37, Feb., pp. 201–205; Mar., pp. 338–341). Water is a satisfactory lubricant for phenolic roll-neck bearings provided it is kept sufficiently cool to maintain its viscosity. The basic principle in spraying on water is to apply it so that it stays on the exposed surface of the journal as long as possible: a number of ways are illustrated by which this may be achieved. Grease and oil may also be used as lubricants with this type of bearing material. Applications of phenolic plastics in thrust blocks and rings, side-bearings, and marine bearings are described, together with the necessary arrangements for keeping sand and scale out of them.—J. P. S.

Industrial Uses of Some Polyether Synthetic Lubricants. W. H. Millett. (Yearly Proceedings of the Association of Iron and Steel Engineers, 1948, pp. 569–578). *See* Journ. I. and S.I., 1949, vol. 161, Jan., p. 66.

Lubrication Equipment Maintenance. R. A. Kraus. (Yearly Proceedings of the Association of Iron and Steel Engineers, 1948, pp. 678–686). *See* Journ. I. and S.I., 1949, vol. 161, Mar., p. 262.

Barium Lubricating Grease. C. J. Boner and G. W. Miller. (Yearly Proceedings of the Association of Iron and Steel Engineers, 1948, pp. 698–702). *See* Journ. I. and S.I., 1949, vol. 161, Mar., p. 262.

WELDING AND FLAME-CUTTING

Effects of Arsenic on the Weldabilities of Steels. H. Sekiguchi, S. Ando, K. Hotta, and Y. Moriwaki. (Welding Journal, 1949, vol. 28, Jan., pp. 53–54). The authors have experimentally determined that, in the gas and arc welding of steels containing arsenic, increasing the arsenic content increases the maximum hardness in the heat-affected zone in low carbon steels, and reduces it in chromium–molybdenum or silicon–manganese–chromium steels; the cracking tendency in these special structural steels is also reduced. They have also examined the forge-welding of such steels and conclude that, on account of the formation of an oxide film, promoted by the presence of arsenic, forge welding becomes difficult, although, with low carbon steels and an arsenic content less than 0·38% it is possible if borax is used as a flux.—J. P. S.

Weldability of Cast Low Alloy Steels. V. T. Malcolm and S. Low. (Welding Journal, 1948, vol. 27, Dec., pp. 1029–1033). The mechanical properties of steels WC1 to WC9, a range of alloy steels suitable for high-temperature service in the cast condition, and their weldability characteristics indicated by bend tests, and the Battelle weldability index which depends on length of crack under certain conditions, are given. The compositions are as follows:

Grade	C, %	Mo, %	Cr, %	V, %	Ti, %	Ni, %
WC1	0·30 max.	0·50	0·10	0·04	0·008	0·23
WC2	0·25 max.	0·50	0·12	0·03	0·01	0·18
WC3	0·19	0·50	0·65	0·03	0·009	0·16
WC3M	0·22	0·50	0·80	0·015	0·05	0·18
WC4	0·23	0·50	0·55	0·03	0·01	0·90
WC5	0·25	0·95	0·75	0·02	0·009	0·80
WC6	0·20	0·50	1·25	0·02	0·011	0·19
WC7	0·20	1·10	0·21	0·02	0·008	0·48
WC8	0·15	1·10	0·10	0·20	0·009	0·20
WC9	0·15	0·95	2·50	0·04	0·010	0·36

Dimensional Changes in Steel—Effect of Welding and Local Heating on Residual Stress. L. C. Bibber. (Welding Journal, 1948, vol. 27, Dec., pp. 1009–1023). The author explains

how heating and cooling cause deformation and how residual stresses are set up. Their effects in the fabrication of large structures such as ships are observed, and methods of construction whereby distortion can be cancelled out or minimized are described.—J. P. S.

The Technology of Electrodes Used for Arc Welding. A. Kleandr. (Svařování, 1948, vol. 8, June, pp. 66–73, July, pp. 93–99; Aug., pp. 109–112). [In Czech]. The influence of the various elements of the electrode coatings, particularly that of oxygen, nitrogen, and hydrogen, is described. The properties of the individual types of coating are outlined and their qualities are compared from the point of view of welding speed and quality of the weld produced. The most important considerations involved in the manufacture of electrodes are explained so as to assist the user to select electrodes suitable for a particular task. Classification tables of electrodes according to American and Czechoslovak Standard Specifications, are included—E. G.

Inert Gas Shielded Arc Spot Welding. F. J. Pilia. (Welding Journal, 1949, vol. 28, Jan., pp. 5–11). A new type of arc spot welder is described. In pistol form, it contains a tungsten electrode situated centrally in, but insulated from, a water-cooled cup which is pressed against the work; on pressing the trigger, argon is admitted round the electrode, and a high frequency current is discharged between it and the work. The welding current then follows along the ionized path, a timer extinguishes it after the correct time cycle, the argon is cut off a few seconds later when the weld is sufficiently cooled; the cycle may then be repeated. As this process is operated from one side only, provided the work is properly supported, it has many applications in the fabrication of articles from sheet up to 0·078 in. in thickness.—J. P. S.

Production Heli-Welding Stainless Steel. B. Blickman and N. Blickman. (Welding Journal, 1948, vol. 27, Nov., pp 945–950). Applications of the heli-arc welding process to stainless steel, in particular to the joining of thin sheets are described; an especial advantage of this process is the small amount of reinforcing, or excess metal on the top of the weld, that it produces. This permits further forming operations if necessary, and demands little finishing. The circumstances in which tungsten, alternatively carbon, electrodes are selected are given, and the use of filler material in the form of wire, strip, or shearings is considered.—J. P. S.

Some Metallurgical Aspects of Austenitic Welds. C. T. Gayley. (Welding Journal, 1949, vol. 28, Jan. pp. 24–30). Tests on deposits of austenitic weld metal indicate that crater cracks, occurring when the arc is broken at the end of a run, can be prevented by filling the crater, and drawing the arc forward along the parent metal before breaking; root cracks may be avoided by the deposition of a heavy first pass with medium to low current. It is found that the grains of austenitic weld metal are laminar, and that the grain boundaries are sources of weakness, some heats from which the core wires are made being subject to intergranular cracking which is very detrimental. Austenitic weld metal shows directional tensile properties; cold working followed by recrystallization increases the tensile strength at the expense of ductility and removes some, but not all of the directionality. J. P. S.

The Economics of Hardfacing. J. J. Barry and A. Muller. (Welding Journal, 1949, vol. 28, Jan., pp. 31–37). The types of service in which hardfacing may be of value are classified as: (1) Severe impact and moderate abrasion *e.g.*, shovel and excavator bucket lips; (2) severe abrasion and moderate impact, *e.g.*, plough shares, cement screw-conveyors; (3) abrasion, impact, and erosion under hot and corrosive conditions, *e.g.*, diesel engine exhaust valves; (4) extreme abrasion, *e.g.*, earth drilling. No one "hard metal" is suitable for all these conditions. For the first type, martensitic air-hardening steels, such as one containing 6% of chromium and 1% of carbon may be employed. For the second type, austenitic or martensitic alloy irons, such as one containing C 4·5%, Cr 30%, Mn 6%, and Si 2%, which will consist of alloy carbides in an austenitic matrix, or 4% C, 4% Mo, and 4% Cr, which gives a martensitic matrix containing alloy carbides. For the third type, choice may be made between a high carbon alloy, such as one having 2·6% C, 3·6% Mn, 32% Cr, 43% Co, 14% W, where abrasion resistance is a desideratum, and a lower carbon alloy, such as 1% C, 30% Cr, 62% Co, 4% W, where impact resistance is desired. Both types have high red hardness and corrosion and erosion resistance. For the fourth type, a structure consisting of hard particles embedded in a tough matrix is used; the hard particles consisting of carbides, nitrides, or borides: tungsten carbide is frequently employed, and an alloy of 1% C, 70% Ni, 15% Cr, 10% Fe (max.) and 3% B forms a hard complex network of boride and complex carbides in a tough solid-solution matrix. Some particular applications of all these hardfacings are given.—J. P. S.

Welding Tests of Nickel-Steel. J. W. Crossett. (Railway Mechanical Engineer, 1948, vol. 122, Nov., p. 111; Railway Engineering Abstracts, 1949, vol. 4, Jan., p. 7). The author records laboratory work done on the welding of nickel-steel boiler plate and gives test results which show that increased strength and toughness may be obtained by the use of welded nickel steel for locomotive boilers. Tests for strain-ageing were among those made. This article is an abridgment of the paper read at the Chicago meeting of the Master Boiler Makers' Association held in September, 1948.

Manganese-Vanadium Steel. T. W. Merrill. (Railway Mechanical Engineer, 1948, vol. 122, Nov., pp. 112–113; Railway Engineering Abstracts, 1949, vol. 4, Jan., p. 7). This is an abridgment of a paper read before a meeting of the Master Boiler Makers' Association held at Chicago in September, 1948. The author considers that manganese–vanadium steel, the properties and weldability of which are described, should be useful to railways as a boiler and firebox material. The welding tests made are described and the benefits arising from the use of this steel are set out.

Steels and Their Edge Preparation. A. J. Raymo. (Railway Mechanical Engineer, 1948, vol. 122, Nov., p. 111; Railway Engineering Abstracts, 1949, vol. 4, Jan., p. 7). In this paper, presented at a meeting of the Master Boiler Makers Association held at Chicago in September, 1948, some account is given of the development work done by the Baldwin Locomotive Works in America to discover the usability of certain types of steel for welded boilers, having regard to economy, suitable types, and methods of edge preparation. The steels favoured are given. The author discusses the question of stress-relieving and expresses the opinion that a boiler could be put in service without prior stress-relieving.

Oxygen Gouging in the Production of Locomotive Boilers. J. Mandaus. (Svařování, 1948, vol. 8, No. 10, pp. 137–141). [In Czech]. The paper gives a brief review of the experience gained with oxygen gouging in the locomotive works of Českomoravská Kolben Daňek. This method has been introduced after the war to lower production costs, it can be used for carbon and alloy steels. Its main application is for gouging the roots of welds, thus replacing pneumatic chipping, but it is also suitable for rough recessing work instead of planing or milling. Speed of work depends on the skill of the operator and accessibility, and it takes 5 to 20 min. to gouge 1 m. of weld. The same work requires 20 to 60 min. with pneumatic tools. Defects are discovered more easily; they cause dark spots to appear in the flame. Cavities are not filled in because the high oxygen pressure does not allow filling in of those cavities by liquid metal. Well over 300 locomotive boilers have been prepared for welding by application of oxygen gouging since the end of the war, and the results appear to be satisfactory.—E. G.

Chrome Steel in Locomotive Firebox Applications. H. L. Miller. (Welding Journal, 1949, vol. 28, Jan., pp. 55–58) Chromium steel (Cr 18%) has shown excellent service in fireboxes, firebox side plates, and siphons in locomotive boilers. Instructions are given for the welding of this material, to itself, to carbon–molybdenum steel, and to open-hearth steel boiler plate.—J. P. S.

Welded Boilers and Alloy Steels. (Railway Mechanical Engineer, 1948, vol. 122, Nov., p. 110; Railway Engineering Abstracts, 1949, vol. 4, Jan., p. 8). The position as regards welded locomotive boilers in the United States is set out. Welded boilers, it is anticipated, may have longer life than riveted ones. No welded locomotive boilers have been constructed of alloy steels. This report was presented at a meeting of the Master Boiler Makers' Association held at Chicago in September, 1948.

Automatic Arc Welding Alloy Steel Plates. H. L. Miller. (Welding Journal, 1948, vol. 27, Nov., pp. 916–920). The author claims that "straight" nickel and molybdenum are, and will continue to be, difficult to manufacture owing to the residual elements introduced by war scrap. Experiments have therefore been carried out on alloy steels containing

manganese and vanadium, nickel and molybdenum, nickel, copper, and molybdenum, and manganese, vanadium, nickel, and molybdenum, in plate form; physical tests on these welded with various electrodes have shown that all the above elements, which tend to form ferritic structures, can be used in alloy steels intended for automatic arc welding. The use of low hydrogen fluxes, and the laying down of small deposits at each pass, and where possible welding in unrestrained conditions, are advocated.—J. P. S.

Alloy Steels for Welded Boilers. R. L. Heath. (Railway Mechanical Engineer, 1948, vol. 122, Nov., pp. 110–111: Railway Engineering Abstracts, 1949, vol. 4, Jan., p. 8). This is a résumé of a paper on the use of alloy steels for welded boilers presented at the meeting of the Master Boiler Makers' Association, held at Chicago in September, 1948.

Hard Facing with Inert-Gas-Arc Welding. K. H. Koopman. (Welding Journal, 1949, vol. 28, Jan., pp. 46–52). The author describes the use of inert-gas-shielded arc methods for hard-facing mild steel, copper, copper-base alloys and stainless steel, and surfacing mild and stainless steels with metals other than hard-facing alloys, such as with bronze and copper, silver, lead, and stainless steel. He concludes that: (*a*) Where dilution of the hard-facing material by the base may occur, it is less with these methods than with the electric arc, though more than with oxy-acetylene methods; (*b*) distortion will be less because the rate of heat input is higher; and (*c*) that good quality deposits are obtainable. A back-hand welding technique with high-frequency stabilized alternating-current for the first layer produces maximum weld hardness and mimimum penetration; straight polarity direct current for subsequent layers, with forehand technique, is excellent for deposition on mild steel. Techniques for the deposition of, and on, other metals are also given.—J. P. S.

Further Studies in Projection Welding. W. F. Hess, J. Childs, and R. F. Underhill, jun. (Welding Journal, 1949, vol. 28, Jan., pp. 15-s–23-s). The projection welding of three steels, AISI 1010 (0·08% C, 0·36% Mn), AISI 1015 (0·15% C, 0·54% Mn), AISI 1020 (0·20% C, 0·45% Mn) in gauges of 0·040 in., 0·062 in. and 0·125 in. has been studied, with various shapes of projection, welding time, and electrode pressure. It is concluded that in 0·040 in. steels the projection diameter should not be less than 0·010 in., but that the exact diameter is not critical above this value, 0·14 in. proving most satisfactory. Most consistent results and better strengths are obtained with a projection height 20% of the diameter. Failures by tearing out of the welded 'plug,' considered the most desirable type of failure, were only observed where fusion had occurred: this was not the case in the 0·040 in. sheet of AISI 1015 and 1020 steels. Where electrodes of different resistances were used it was found that less welding current was required with higher resistance electrodes, probably because the heat loss from the weld to the electrode was less. Multiple welds had less strength per weld than single welds, the difference being greater in the higher carbon steels.—J. P. S.

What is the Best Way to Repair Cast Iron? K. H. Koopman. (Welding Journal, 1948, vol. 27, Nov., pp. 951–954). The differences between fusion welding and 'braze' welding (*i.e.*, bronze welding or brazing with the use of a filter rod and the oxy-acetylene torch) are described, and indications given of the suitable circumstances in which to use each process. A table illustrates appearance, chip, blowpipe, and spark tests for distinguishing between white, grey, and malleable iron.—J. P. S.

Welding in Steel Mill Maintenance. L. P. Elly. (Welding Journal, 1949, vol. 28, Jan., pp. 38–45). The use of various types of welding, such as submerged-arc welding, bronze welding and thermit welding, as well as the conventional arc or oxy-acetylene methods, in repairing and building-up broken or worn steel mill equipment, is described and illustrated.—J. P. S.

Semi-Automatic Oxygen Flame Cutter for Steels 100 to 300 mm. Thick. (Avtogennoe Delo, 1948, No. 9, pp. 26–27: Svařování, 1948, No. 10, pp. 145–146). [In Czech]. The Soviet Vniiavtogen Institute has developed a semi-automatic flame cutter for cutting steels of thicknesses between 100 and 300 mm. It was specified that the deviation from the desired cutting angle should not exceed ± 1°, the deviation of the cutting line from a straight line should not be larger than ± 0·1% of the length cut, the depth of cutting irregularities should be less than 0·5% of the thickness of the cut steel, and the radius of curvature of the edges should be smaller than 4 mm. in the first 50-mm. portion of the cut and below 0·5% of the thickness of the plate for the remaining length. The developed cutter is suitable for use on 110–124 V. it weighs 52 kg. including the burner, but excluding the guide rails; its size in plan (without guide rails) is 600 × 670 mm., the deflection angle between burner and metal surface can be varied between 0 and 35°. The stability of the cutting device on the guides is of great importance, and a 1-mm. slag particle on one of the rails results in a cutting irregularity of about 1·5 mm., the rails must therefore be protected by shields. Experience has proved that there is a definite optimum value for the pressure of the oxygen and the cutting speed in each case, and that it is inadvisable to increase the pressure above that value; if the pressure is increased the cutting speed will also be higher, but the quality of the cut will suffer. The bottom of the first section of the part to be cut should be preheated, and the time required for this purpose is 60 to 90 sec. Numerous practical tests showed that this cutter complies with the specified requirements.—E. G.

Cutting Stainless Steel and Other Oxidation-Resistant Alloys. R. Priess. (Southern Power and Industry, 1948, vol. 66, No. 1, pp. 62–64: [Abstract] British Abstracts, B.I. 1948, Oct., col. 563). Three new processes have been developed for flame-cutting of oxidation-resistant alloys. The Linde-Air process uses a non-hygroscopic, non-toxic, finely divided, iron-rich powder which is fed into the stream of oxygen from outside the preheat flame. The powder burns in the oxygen flame, creating a high-temperature reaction, and the products of combustion provide the necessary fluxing action. Powder flows are ¼–½ lb./min. and cutting speeds are comparable with those with carbon steel. The Air Reduction process uses an oxide-eliminating powder flux which is carried by the cutting oxygen through the tip to the point of the cut. The Arcos high-temperature reaction is obtained by the use of an electric arc and a stream of oxygen, using a tubular mild-steel-coated rod, a special holder, oxygen supply, and either an A.C. or D.C. welder. None of these methods influences the corrosion-resistance, because of the subsequent processing.

MACHINING AND MACHINABILITY

Friction-Sawing Facts and Figures. H. J. Chamberland. (Engineers' Digest (London), 1949, vol. 10, Jan., pp. 23–25). The applications and process of friction-sawing are discussed. Cutting speeds for various steels are tabulated.—R. F. F.

Nature and Detection of Grinding Burn in Steel. L. P. Tarasov and C. O. Lundberg. (American Society for Metals, Oct. 1948, Preprint No. 44). This paper is concerned with the microstructural and hardness changes that may occur in hardened steel parts when they are burned in grinding, and with their response to macroetching. Most of the results were obtained on high-speed and oil-hardened tool steels, but enough work was done with other steels to show that the same conclusions can be expected to apply to hardened steels generally. Hardness data are presented showing that, with good grinding practice, hardened steel can be ground without appreciable surface softening, and that burn can be completely eliminated if necessary. It is also shown that grinding can work-harden the surface of hard steel. Sensitive, non-destructive macroetching techniques are described for detecting hardness changes (and sometimes high stresses) in ground surfaces; these should be useful for inspection purposes. J. C. R.

How to Solve Milling Problems with Carbide Tools. (Technique Moderne, 1948, Apr. 1 and 15, pp. 119–123; June 1 and 15, pp. 182–183: [Abstract] Revue Universelle des Mines, 1948, Series 9, vol. 4, Dec., p. 724).

Superfinish. A. Mauzin. (Revue Générale de Mécanique, 1949, vol. 33, Jan., pp. 5–10). The experiments described in this article were carried out to study the process of superfinishing, the qualities required of abrasive wheels and operating conditions.—R. F. F.

Physical Analysis of Mechanical Superfinishing. J. Digard de Cuissart. (Revue Générale de Mécanique, 1949, vol. 33, Feb., pp. 71–80). The author studies the mechanism of superfinishing, discusses the properties of abrasive stones, the characteristics of superfinishing machines, and their effect on the surface obtained after superfinishing.—R. F. F.

Standardization of Surface Quality. H. von Weingraber. (Die Technik, 1948, vol. 3, Oct., pp. 417–423). It is shown that the standardization of the quality of metal surfaces is

long overdue. The relation between surface geometry and methods of production is discussed. Methods of measuring surface contours are described, and proposals for a system of classification are made.—R. A. R.

Surface Roughness Measurements. H. Naumann. (Metallwirtschaft, 1948, Oct., pp. 326–330). Several methods of measuring surface roughness are briefly reviewed.—R. A. R.

The Presentation and Measurement of Surface Roughness with the Transmission Electron Microscope. R. Seeliger. (Metalloberfläche, 1949, vol. 3, Jan., pp. 9–15). Methods of obtaining surface replicas for use in the electron microscope to make measurements of surface roughness are described.

R. A. R.

Methods of Measurement in Machinability Research. O. E. Svahn. (Instruments and Measurements Conference, Stockholm, 1947, Transactions, pp. 123–128). Methods of measuring cutting forces, tool vibration and temperature, punching forces, and of checking tool angles are described. In particular details are given of the author's cutting force pick-up. In this the tool is held in a fixture which, through balls or pins, rests on the diaphragms of three pressure pick-ups, one under the tool taking up the tangential force, one at the side of the tool taking up the feed-force or axial force, and one at the rear end of the tool taking up the radial force. The movements of the diaphragms are measured electrically.

R. A. R.

Metal Cutting Fluids. E. L. H. Bastian. (Iron Age, 1949, vol. 163, Feb. 10, pp. 60–67). The functions of cutting fluids, in cooling the work, reducing friction between tool and chip and preventing the building-up of metal on the tool point, are described. Distinction is then drawn between cutting oils used neat, and water-soluble oils or compounds, and between active cutting oils the chemical reactivity of which affects the cutting process, and inactive ones which have no such reactivity. Various types of cutting fluid in each of these categories are discussed, and the characteristics of the machining operations which decide the kind of fluid to be employed are outlined. A table indicates the type of cutting fluid for a number of operations on a range of ferrous and non-ferrous metals and alloys.—J. P. S.

CLEANING AND PICKLING

Modernization of the Pickle Room. (Finish, 1948, vol. 5, No. 10, pp. 20–21: American Ceramic Abstracts, 1949, Feb. 1, p. 36). The new pickle room of the Wolverine Enameling Co., where cold muriatic acid is used for pickling, is described.

Relationship of the Chemical Engineering Industry to the Iron and Steel Industry. B. N. Reavell and G. H. Black. (XIth International Congress on Pure and Applied Chemistry, 1947: [Abstract] British Abstracts, B.I, 1948, Sept., col. 430). Chemical engineering problems encountered in the iron and steel industry are fume extraction and pickling by the old method, and by the newer electrolytic method. The possibilities of recovery of sulphuric acid from ferrous sulphate solutions are outlined.

Production of Sheet Steel at Irvin Works Modernized. (Blast Furnace and Steel Plant, 1948, vol. 36, Dec., pp. 1465–1469). The Irvin Works of Carnegie-Illinois Steel Corporation has recently installed an additional slab-heating furnace of a capacity of 105 tons/hr. of slabs 6–8 in. thick by 8–10 in. long, new conveyors and shearing equipment in the line for pickling 80-in. sheet, an entirely new continuous pickler for 56-in. sheet, new sheet-coil annealing furnaces and modifications to flat sheet annealing furnaces, and new sheet shearing and slitting gear.—J. P. S.

Behaviour of Hydrogen in Steel during and after Immersion in Acid. L. S. Darken and R. P. Smith. (Corrosion, 1949, vol. 5, Jan., pp. 1–16). Experiments to determine the permeability of steel to hydrogen, the saturation concentration, and the diffusivity of hydrogen in steel are described. The principal findings were: (1) The amount of hydrogen absorbed during acid pickling is related to the dimensions of the steel in the manner to be expected from the laws of diffusion; (2) the amount absorbed increases practically linearly with the square root of time up to a saturation value; (3) the saturation value is very small for hot-rolled steel, but it increases linearly with cold work; (4) the saturation value is slightly decreased by long-time annealing of cold worked steels, and this annealing considerably decreased the rate of hydrogen absorption; and (5) increasing the *p*H value of the pickling solution has a small effect on the saturation value but a more pronounced effect on the rate of absorption.

R. A. R.

Evaluation of Pickling Inhibitors from the Standpoint of Hydrogen Embrittlement. C. A. Zapffe and M. Eleanor Haslem. (Wire and Wire Products, 1948, vol. 23, Oct., pp. 933–939; Nov., pp. 1048–1053, 1080–1082; Dec., pp. 1126–1130, 1172–1175). Part I of this series reports investigations of fourteen pickling inhibitors to study their effect on hydrogen embrittlement when stainless steel is pickled in sulphuric acid. Eleven of these caused an increase in the embrittlement, two had no effect, and one caused a slight decrease. Two new inhibitors, known as Rodine X-1 and Rodine X-2, have been developed, the addition of which causes less embrittlement in stainless steel than occurs after pickling in uninhibited acid. Part II reports the results of tests with 15 inhibitors when pickling mild steel with sulphuric acid. Seven failed to prevent embrittlement regardless of the concentration; four prevented embrittlement when added in concentrations greater than those recommended for commercial use; four prevented embrittlement when added in normal concentrations and one of these effectively prevented embrittlement of stainless as well as mild steel.

Part III describes investigations of the inhibitors previously studied, but this time the mild steel and stainless steel specimens were made cathodes. Of the 16 reagents studied 11 caused a marked increase in the embrittlement of both mild steel and stainless steel. No inhibitor prevented the embrittlement of stainless steel: some increased it and others had no effect.—R. A. R.

Sodium Hydride Descaling of Stainless and Clad Steels. J. S. Morris. (Yearly Proceedings of the Association of Iron and Steel Engineers, 1948, pp. 827–835: Iron and Steel Engineer, 1948, vol. 25, Nov., pp. 71–79, 84–86). A detailed description is given of the sodium hydride descaling plant of the Lukens Steel Company, with details of the construction of the furnaces and tanks.—J. C. R.

Manual Pickling Operation. A. M. Langbein. (Porcelain Enamel Institute: Industrial Heating, 1948, vol. 15, Dec., pp. 2158–2161). Data are presented on the sizes of tanks, operating conditions, and costs of the American Stove Company's pickling plant for preparing steel sheets for enamelling.

R. A. R.

New Technique in Conditioning Stainless Steel. A. E. Hamilton, jun. (Yearly Proceedings of the Association of Iron and Steel Engineers, 1948, pp. 146–155). *See* Journ. I. and S.I., 1948, vol. 159, June, p. 223.

Continuous Flow Internal Pipe Pickling. (Yearly Proceedings of the Association of Iron and Steel Engineers, 1948, pp. 584–585). *See* Journ. I. and S.I., 1949, vol. 161, Jan., p. 67.

Immersion Gas Tube Process of Descaling Bar Steel. G. D. Muschlitz. (Yearly Proceedings of the Association of Iron and Steel Engineers, 1948, pp. 836–842: Iron and Steel Engineer, 1948, vol. 25, Nov., pp. 80–86). The author describes the Hooker process for descaling bars of stainless steel or high alloy steel prior to cold drawing. A typical treating cycle consists of the following steps: Annealing; fifteen-minute immersion in the salt bath; quick dip into the cold water tank; five-minute immersion in the hydrochloric acid tank; high pressure water rinse; five-minute immersion in the nitric acid tank; high pressure water rinse; five minute immersion in the cleaning solution tank; immersion in the lime tank for lubrication preparatory to cold drawing; and cold drawing. This process is in operation at the works of the Midvale Company and the practice and equipment at that plant are described. Compared with the sodium hydride descaling method, the Hooker process has a much higher operating temperature, the baths are smaller, and gas-fired immersion tubes are used for heating.—J. C. R.

Improved Electrolytes for the Anodic Polishing of Certain Metals. P. A. Jacquet. (Société Française de Metallurgie: Sheet Metal Industries, 1949, vol. 26, Mar., pp. 577–584, 593). Improved solutions and procedures for the electrolytic polishing of metals are described.—R. A. R.

The Reduction of Polishing Costs. S. Wernick. (Journal of the Electrodepositors' Technical Society, 1948, vol. 23, pp. 215–239). Modern methods of polishing metal parts are reviewed with special reference to directions in which economies can be effected so that polishing costs may be reduced.

The Mechanism of Degreasing Metals. S. Spring and L. F. Peale. (Metalloberfläche, 1948, vol. 2, Oct., pp. 209–212). *See* Journ. I. and S.I., 1947, vol. 156, May, p. 137.

Paste for Removing Rust. Mason. (Verniciatura e Decorazione, 1948, vol. 3, No. 5, p. 2 : Railway Engineering Abstracts, 1949, vol. 4, Mar., p. 71). A paste composed of H_3PO_4 2 (parts by wt.), oxalic acid 2, glycerol 1, and SiO_2 5 is spread on the rusty iron article and left for 20 min. It is then washed off with water, leaving the metal perfectly free from rust.

PROTECTIVE COATINGS

Metal Coatings. J. Hoffmann. (Metallwirtschaft, 1948, Sept., pp. 295–297 ; Oct., pp. 330–332). This review covers methods of cladding steel with other metals by compound casting, pressure welding in a protective atmosphere, and welding by rolling. The properties and applications of the composite materials obtained are discussed.

Modern Metering Instruments for Metallurgical Science and Practice. J. Kesper. (Metallwirtschaft, 1948, Nov., pp. 371–372). Descriptions are given of the Poggendorf and the Pehavi instruments for measuring the *p*H value of plating solutions.—R. A. R.

Defects in Plating Solutions and Their Remedies. P. Berger. (Journal of the Electrodepositors' Technical Society, 1945–46, vol. 21, pp. 195–209). *See* Journ. I. and S.I., 1947, vol. 157, Nov., p. 473.

The Chromium Impregnation Process. F. Steinberg. (Archiv für Metallkunde, 1948, vol. 2, No. 8, pp. 253–254). The characteristics of chromium-impregnated steel are discussed. With this gaseous process the surface of the steel contains 35% of chromium which diminishes to 13% at a depth of about 0·1 mm. The principal advantage is the small amount of chromium consumed, namely 160 to 180 g./sq. m. of surface.—R. A. R.

The Effect of Heat on Electro-Deposited Chromium. R. Graham, K. R. Williams, and R. W. Wilson. (Engineering. 1949, vol. 167, Mar. 18, pp. 241–243 ; Mar. 25, pp. 265–267). An investigation was undertaken to study the properties of electrodeposited chromium in the 100–450° C. range of temperature, with special reference to the effect of heat and friction on the wear resistance. The hardness decreases linearly with rise in temperature ; after heating and cooling there is a small but significant drop in hardness, the drop increasing with the number of heating and cooling cycles. On first heating chromium it expanded in a normal manner, the coefficient of linear expansion being $6{\cdot}2 \times 10^{-6}$ per degree between 20° and 300° C. The deposit was not heated above 350° C., and in four subsequent determinations on the same piece of chromium, the expansion recorded up to 300° C. was zero ; at this temperature a contraction was recorded which seemed, qualitatively, to decrease in magnitude with each subsequent heating. Chromium plating less than 0·0003 in. thick is not suitable for wear resistance owing to flaking under load. Snavely's results are used as a basis for the theory covering the stress in chromium deposits and for the crack-producing contraction. Chromium is deposited, if not completely, then to a considerable extent, as chromium hydride and the hydrogen in the coating is driven off by heating, practically all of it being evolved at 300° C. Trouble due to contraction may be experienced down to 80° C. where hydrogen is first evolved : this is borne out by wear tests in which scoring was found at 100° C. after 6 hr. but not after 1 hr. The procedure for obtaining a wear-resisting chromium coating for cylinder liners is described.

An appendix describes the determination of impurity in electrolytic chromium. Two results obtained were 2·4% and 3·3% of impurity which had a composition approaching that of one of the oxides CrO_3 or Cr_2O_3. It is postulated that it was a decomposition product arising in the electrolysis of the CrO_3 plating bath.—R. A. R.

Bright Chromium Plating. (Electroplating 1949, vol. 2, Jan., pp. 6–11). Details of operating practice in bright chromium plating are outlined. A tabulated summary is given listing defects, their possible causes, and suggested remedies.—J. C. R.

Outplating at the Verichrome Plating Co., Walsall, England. (Electroplating, 1948, vol. 1, Sept., pp. 574–583 ; Oct., pp. 638–645 ; Nov., pp. 718–719, 723). An illustrated account is given of the plant of the Verichrome Co., Ltd., and the practice at its works, including zinc, cadmium, tin, dull and bright nickel, chromium and decorative brass plating ; zinc, cadmium, tin and chromium barrel plating ; and oxidized copper and bronze finishes.—J. C. R.

Physical Properties of Electrodeposited Metals. I. Nickel. (Plating, 1948, vol. 35, Dec., pp. 1228–1239). Extensive test data are presented which show how changes in the plating conditions affect the following properties of nickel coatings : Hardness, tensile strength ; elongation, Young's modulus, fatigue strength, density, electrical resistivity, magnetic properties. The effects of coating thickness, composition, and heat-treatment are also noted.

Determination of Impurities in Electroplating Solutions. X. Traces of Aluminium in Nickel Plating Baths. (Plating, 1948, vol. 35, Oct., pp. 1019–1022, 1054). The method which is proposed for determining aluminium in nickelplating baths is a standard colorimetric procedure with aluminium reagent. Iron is first removed by extraction of iron cupferride with amyl acetate. The large amount of nickel present is removed by complexing with sodium diethyl dithiocarbamate followed by extraction with chloroform. The colour of the aluminon lake is developed by adding aluminon and a buffer to adjust the *p*H to 5·5. After standing 30 min., the transmittancy of the lake is measured with a Klett-Sommerson colorimeter using a No. 54 green filter.—R. A. R.

Large Scale Electroplating Operations Combine Automatic and Bulk Processing. F. C. Fischrupp. (Steel, 1949, vol. 124, Apr. 4, pp. 102–106). In telephone apparatus manufactured by the Western Electric Co., Chicago, considerable amounts of zinc and of nickel plus chromium plating are used, for corrosion and wear resistance, not for appearance. Automatic platers are used where the quantity justifies it, and "rotobarrel" plating in other conditions. For measuring zinc-layer thickness, the Magne-gage is used or, where the size or shape of the pieces prevent this, a chemical stripping method : the thickness of nickel and chromium deposits is measured by a cross-sectioning technique.—J. P. S.

Approximate Tables for Drawing Up Hot-Dip Galvanizing Pot Specifications. (Industrial Finishing, 1948, Nov., pp. 221–222). Methods of calculating the volume and weight of zinc and the required heat input per pound of steel to be galvanized are explained and data are presented based on 20, 15, and 10 lb. of molten zinc per pound of production.—R. A. R.

Electrolytic Tinplating at the Republic Steel Corporation. (Electroplating, 1949, vol. 2, Feb., pp. 83–88). An illustrated description of electrolytic tinplating practice at the Republic Steel Corporation, Niles, Ohio, U.S.A. is presented. The plant handles tinplate from 16 to 38 in. wide and 0·007 to 0·03 in. thick and gives a tin coating thickness of 0·00003 in. The normal operating speed is 500 ft./min.—J. C. R.

Porosity in Metal Coatings. Formation and Assessment : Note on Work in Progress at B.I.S.R.A. J. Pearson. (Electrodepositors' Technical Society, Jan. 17, 1949 : Electroplating, 1949, vol. 2, Feb., pp. 102–106). The author discusses the occurrence and causes of porosity, particularly in electrolytic and hot-dipped tinplate. Methods of pore counting and assessing areas of discontinuities are briefly summarized, and reference is made to the uniformity of tin coatings and the assessment of porosity by the cathodic suppression of current.—J. C. R.

Continuous Gas-Radiant Fusion for Electrolytic Tinning. C. E. Cunningham. (Yearly Proceedings of the Association of Iron and Steel Engineers, 1948, pp. 590–597). *See* Journ. I. and S.I., 1949, vol. 161, Jan., p. 69.

Progress in Developing Palm Oil Replacements. E. L. H. Bastian. (Steel, 1949, vol. 124, Feb. 14, pp. 84–88, 124). A substitute for palm oil for use in the tinning of sheet has been developed by the Shell Oil Co. Two oils are used, one as the initial charge or fill, the other added during working to make up losses due to drag out and volatilization. The yield and quality of tinplate produced during a test run with these oils were satisfactory. Mineral-base oils have also been tested in the grinding of stainless steel sheets where palm oil was formerly employed and considerable savings were found.—J. P. S.

Aluminium Coating Steel. C. B. Ulshafer. (Wire and Wire Products, 1948, vol. 23, Dec., pp. 1124–1125). Methods of coating steel wire with aluminium are briefly reviewed with special reference to those of C. G. Fink, T. Sendzimir, and M. Alferieff, all of which are based on the removal of oxides

from the surface of the steel by reduction in hydrogen, and leading the wire into a bath of molten aluminium without permitting any contact with air.—R. A. R.

Applying Industrial Finishes. G. J. Cavanaugh. (Steel, 1949, vol. 124, Mar. 21, pp. 106–108). Types of organic finishes and methods of applying them are briefly described.—J. P. S.

"Araldit" Resin Coating for Protecting Metal Surfaces. G. H. Ott. (Schweizer Archiv, 1949, vol. 15, Jan., pp. 23–31). **Some Properties and Applications of a New Coating Lacquer.** G. H. Ott. (Sheet Metal Industries, 1949, vol. 26, Feb., pp. 367–370, 381). Details are given of the constitution, properties, and applications of "Araldit," a synthetic coating material of the aethoxylin-resin type, developed by CIBA A.G., Basle. It has a very high resistance to the attack of acids and alkalis and is used for steel sheets and light metals, particularly for food cans and pharmaceutical preparations.—R. A. R.

A Cell for the Measurement of the Water Permeability of Lacquer Paint Coatings. A. Ya. Korolev. (Zavodskaya Laboratoriya, 1949, vol. 15, Jan., pp. 120–122). [In Russian]. The permeability to water of protective coatings is measured, by the method described, by following the loss in weight of a special cell the base of which is a 40μ membrane of the coating material; the cell contains water and is placed in a desiccator.—S. K.

The Phosphatization Layer—Its Appearance, Its Thickness. E. Jaudon. (Métaux et Corrosion, 1947, vol. 22, July, pp. 121–124). An illustrated description is given of microscopical examinations of phosphatized layers and methods of determining their thickness.—J. C. R.

The Pre-Treatment of Metal Surfaces by Phosphating. M. A. Streicher. (Industrial Finishing, 1948, Oct., pp. 158–162; Nov., pp. 233–236). *See* Journ. I. and S.I., 1949, vol. 161, Jan., p. 69.

New Developments Widen Metallising Uses. J. E. Wakefield. (Iron Age, 1949, vol. 163, Mar. 17, pp. 81–85). Three new materials for metal spraying are described. One is a new aluminium–iron–bronze of wear resistance and strength approaching those of cast iron and the second a molybdenum alloy which bonds readily, on spraying on any clean smooth metal surface except that of copper, and forms a basis for the spraying of other metals. The third (composition not given) is a powder mixed with a plastic into the form of a rod or wire, sprayed on and then fused with a torch or in a furnace; a type of hard-facing is thus produced. The value of sprayed metals in reducing friction and providing good lubricating properties is also noted.—J. P. S.

Metallising in Relation to Foundry Practice. J. B. Stiles. (Proceedings of the Institute of British Foundrymen, 1947–48, vol. 51, pp. A202–A210). *See* Journ. I. and S.I., 1949, vol. 161, Jan., p. 70.

Metallizing as a Production Process. K. B. Smith. (Welding Journal, 1949, vol. 27, Nov., pp. 941–944). Some applications of metal-spraying are given, such as the restoration of worn bearings and journals and the rectification of undersized machined parts. An essential factor is the preparation of the surface to the correct mechanical roughness; where parts are so designed that this cannot be quickly and cheaply attained by, *e.g.*, abrasive blasting, the savings produced by metallizing may be vitiated.—J. P. S.

Sprayed Metal Coatings—Their Structure, Properties and Uses. J. E. Wakefield. (Metal Progress, 1948, vol. 54, Dec., pp. 827–832). The porosity of sprayed metal coatings is discussed. For protection against corrosion, porosity limits metal spraying to metals that are anodic to the base: thus zinc and aluminium protect iron and steel in spite of slight porosity. Metals that are cathodic to iron and steel are not suitable for this purpose unless the pores in the sprayed coating can be sealed. Porosity may be advantageous in mechanical applications for surfaces that must be lubricated as (*a*) irregularities in the surface provide spaces for metal particles that are dislodged during the wearing-in process, and (*b*) the pores absorb and hold appreciable quantities of oil. This is indicated in a table comparing the coefficients of friction of sprayed steel and hardened steel using plain oil and colloidal-graphited oil for various loads and speeds. Details are given of the tensile properties and Rockwell hardness numbers of some sprayed metals. Finally, the advantages and limitations of metal spraying are considered, and its fields of application—electrical, protective, and mechanical—are reviewed.—J. C. R.

The Properties of Sprayed Metal Coatings. H. Reininger. (Metalloberfläche, 1948, vol. 2, Sept., pp. 185–192: Oct., pp. 213–218). This is a comprehensive review of the properties of sprayed metal coatings. The density of the coating is less than that of the same metal in the cast or forged state. The presence of oxides increases the density of sprayed light metals and lowers that of heavy metals. The electrical conductivity is much less than that of the metal in wire form. The microporosity reduces the hardness of the metal as such, but the oxide content may result in greater hardness than that of the metal as cast. With steel the effect of oxides may be so great that it cannot be machined with ordinary tools. All the properties of the coating are strongly affected by the type of pistol and the technique used.—R. A. R.

Preparing Special Steels for Enameling. F. R. Porter. (Porcelain Enamel Institute: Steel, 1949, vol. 124, Feb., 14, pp. 90–92). The thinner an enamel coat the better will it resist warpage and chipping. Enamelling sheets containing titanium are most suitable for covering with titanium oxide enamels, and the enamel coating can be reduced to 0·0065 in. thick compared with the customary 0·015-in. layer of conventional enamels on "enamelling iron." These titanium oxide enamels will cover satisfactorily all types of weld used in construction. Thorough cleaning of the sheets is essential, and a nickel flash is required for enamel adherence. A modified falling-weight test can be used to test the adherence.

J. P. S.

Base Metal Selection Important to Successful Porcelain Enameling. F. R. Porter. (Materials and Methods, 1949, vol. 29, Feb., pp. 62–64). The characteristics from the point of view of enamelling of mild steel, enamelling iron, titanium enamelling steel, and coated mild steel are discussed. Coated mild steel carries a layer of nickel about 0·000015 in. thick, and is intended for enamels to be stoved at 1300° F. instead of the more usual temperature of 1500° F. Its resistance to sagging is poor at 1600° F., but good at 1300° F.—R. A. R.

Effect of Refractory Mill Additions on Molten Apparent Viscosities of Steel Ground-Coat Enamels. B. D. Bruce and A. V. Sharon. (Journal of the American Ceramic Society, 1949, vol. 32, Feb. 1, pp. 41–45). The rotating cylinder method for the determination of viscosity at high temperatures was used in this investigation. Although this method has been largely confined to the glass industry, it has some advantages over the "button method" for determining flow properties, which is in common use in the enamel industry. The apparatus and procedure used are described. Three frits in common use were blended and various amounts of refractories were added to this blend. Results show that silica imparted the greatest increase in viscosity. The other two refractories tested, feldspar and nepheline syenite, had about the same effect on the viscosity up to 9% addition, but at 12%, feldspar caused greater viscosity than the nepheline syenite.—R. F. F.

Ford Oven Installation Combines Controlled Convection and Infra-Red Heating. (Industrial Heating, 1948, vol. 15, Dec., pp. 2154–2156). Brief particulars are given of the nine drying ovens installed at two Ford assembly plants for the rapid drying of car bodies, fenders, hoods, and wheels after the priming and first colour coats have been applied. Controlled combined convection and infra-red heatings are applied to bring the surface temperature to 325° F. in 3 min. and to maintain this for from 6 to 12 min.—R. A. R.

Comparison of Various Antimonates as Opacifiers in Dry-Process Enamels for Cast Iron. W. F. Coffeen. (Journal of the American Ceramic Society, 1949, vol. 32, Jan. 1, pp. 10–14). Various metallic antimonates were compared with sodium antimonate to determine their suitability as substitutes for sodium antimonate, and the effect on the colour and opacity of antimony oxide enamels when antimonates other than sodium antimonates are formed. All antimonates were compared in high-lead acid-resisting enamels and some were also compared in leadless and lead-bearing non-acid-resisting enamels. The data obtained in these tests show that: (1) Although all antimonates contain pentavalent antimony, their effect on enamel colour and their opacifying ability is not equal; (2) barium antimonate produces results approaching those produced by sodium antimonate for opacifying ability and effect on colour for all three types of enamel, and will also produce enamels of improved acid resistance; (3) the adverse effects of many of the antimonates on enamel colour

and opacity offer an explanation of the inconsistent results often obtained when the dry-process enamels are made by using antimony oxide.—R. F. F.

Control Procedure and Its Administration in Enamel Plants. M. M. Reagan. (Enamelist, 1948, vol. 25, No. 8, pp. 9–12, 75 : American Ceramic Abstracts, 1949, Jan. 1, p. 5). Quality control, a statistical record, and analysis of day-to-day production, will improve product quality, decrease rejections, show economies in labour and material, decrease inspection reports, improve producer–consumer relations, and provide economical specifications for high-speed production. The control programme can include new materials, processes, product quality, and a system of costs and records. Suggested points of control are given for raw materials, pickling, milling, and tests for the final product. The discussion of cost records details the data required to provide the cost of enamelling per square foot.

Relation of Firing Temperature to Steel Sag on Enameling Iron and Cold-Rolled Steel. B. D. Bruce. (Better Enameling, 1948, vol. 19, No. 8, p. 8 : American Ceramic Abstracts, 1949, Jan. 1, p. 6). Cold-rolled steel, because of lack of strength at ordinary enamelling temperatures (1550° F.), shows far more sag when porcelain enamelled than does enamelling iron. The use of the new low temperature (1300° F.) enamels obviates this difficulty. In fact, cold-rolled steel shows less sag when enamelled at 1300° F. than does enamelling iron enamelled at 1550° F.

Relation of Firing Temperature to Possible Enameling Defects Resulting from Steel Gas Evolution. B. D. Bruce. (Better Enameling, 1948, vol. 19, No. 9, p. 18 : American Ceramic Abstracts, 1949, vol. 32, Jan. 1, p. 6). The evolution of hydrogen occluded in steel is greatly enhanced above 1450° F., and heating above the critical temperature favours the liberation of gases from other sources in the steel–enamel system. Enamels maturing at 1300° F., therefore, will probably not be as subject to such defects as blisters and pinholes as those maturing at 1500° F., and above.

Enameling Auto Window Moldings by Electrostatic Spray. T. S. Blair. (Iron Age, 1949, vol. 163, Feb. 10, pp. 75–77). The use of electrostatic spraying in the manufacture of automobile window frames is described and illustrated. J. P. S.

Drying Paints with Gas-Heated Radiation Driers. C. A. Landfermann. (Metalloberfläche, 1949, vol. 3, Jan., pp. 14–16). The advantages and limitations of gas-heated paint drying are discussed and compared with those of electric equipment.—R. A. R.

POWDER METALLURGY

The Most Important Commercial Processes for Producing Iron Powders and Their Influence on the Properties of the Powders. H. Bernstorff. (Archiv für Metallkunde, 1948, vol. 2, No. 9, pp. 289–295). The following processes for the production of iron powder for powder metallurgy are described : (1) The Hametag process ; (2) the D.P.G. centrifugal process ; (3) the D.P.G. pressure jet process ; (4) the Vogt process ; and (5) the carbonyl process. The economy of these processes and the availability of the raw materials are dealt with. Data on particle size, shape, and bulk density of the powders, and bend-test results on specimens made by powder metallurgy are presented. The bend test is shown to be a more sensitive test than bulk-density determinations for ascertaining the suitability of powders.—R. A. R.

Particle Size Determination by Sedimentation, and Its Application in Powder Metallurgy. N. Thien-Chi. (Annales de Radio-électricité, 1948, vol. 3, No. 12, p. 115 : British Ceramic Abstracts, 1948, Nov., p. 374A). The pipette and balance forms of sedimentation analysis were used. For the first, the Andreasen pipette was modified ; for the second, an automatically recording balance was developed on the basis of an ordinary chemical balance of the chainomatic type.

Pressure Distribution in Compacting Metal Powders. P. Duwez and L. Zwell. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 2515 : Journal of Metals, 1949, vol. 1, Feb., Section 3, pp. 137–144). In this investigation the pressure at various points on the sides and bottom of a die compressing metal powder was measured by means of small piston dynamometers and resistance strain gauges. Pressure distribution within the compact was determined by an indirect method based on the relation between the local density within the compact and the compacting pressure that would be required to obtain such a density. This latter method was used to correlate results obtained by the first method with those of other investigators. Pressure on the walls in a cylindrical die varied almost linearly from bottom to the top of the compact. On the fixed piston the pressure decreased from the centre to the periphery. There was a radial distribution of pressure at the top and bottom of the compacts. Near the moving piston, the pressure was high at the outer edge and decreased towards the centre. At the fixed piston, the pressure was higher at the centre of the compact and decreased towards the outer edge. Radial distribution at the top was always present, but at the base it decreased with increasing thickness of compact until it vanished.

Lead-Grid Study of Metal Powder Compaction. R. Kamm, M. A. Steinberg, and J. Wulff. (American Institute of Mining and Metallurgical Engineers Technical Publication No. 2487 : Metals Technology, 1948, vol. 15, Dec.). The authors have previously described the lead-grid method for exploring the distribution of density and strain within metal powder compacts (*see* Journ. I. and S.I., 1947, vol. 156, Aug., p. 584). Circular-hole (instead of rectangular) grids have now been used as they could be more accurately made and, when deformed within the powder, could be more readily measured and analysed. In this paper the effect of lubrication, compact height, pressure, speed of pressing, and vacuum pressing are discussed. Adequate lubrication was essential to achieve maximum average density and uniformity of density distribution ; it was more important in overcoming die-wall friction than inter-particle friction. For relatively hard powders, such as nickel and iron, the speed of pressing, pressing from both sides, and vacuum pressing appeared relatively unimportant as compared to adequate die-wall lubrication in obtaining maximum density and uniformity of density. Hard powders were not measurably affected by increased speed of pressing but soft powders gave compacts of decreasing density. The results of vacuum pressing indicated that not much was to be gained by this technique when applied to hard powders.—J. C. R.

Self-Diffusion in Sintering of Metallic Particles. G. C. Kuczynski. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 2528 : Journal of Metals, 1949, vol. 1, Feb., Section 3, pp. 169–178). The author has studied the mechanism of the process whereby two or more particles bond together with application of heat at temperatures below the melting point of any of the particles. After reviewing the work of other investigators, he considers whether this sintering process is due to plastic flow, evaporation and condensation, volume diffusion or surface diffusion. He gives an account of experiments with copper and silver powders. Results indicated that sintering at atmospheric pressure is predominantly due to volume diffusion for large particles and higher temperatures. It was possible to measure the coefficient of self-diffusion without radio-active tracers, measuring only the rate of growth of the interface between round particles and a metallic block. At the beginning of sintering surface diffusion was operative. J. C. R.

Methods of Pressing for Producing Sintered Steel Parts of Complicated Shape. H. Silbereisen. (Archiv für Metallkunde, 1948, vol. 2, No. 9, pp. 301–305). The design of presses, dies, and punches for the powder-metallurgy process is discussed, the disadvantages of older methods are pointed out, and a new process and machines for making parts with irregular contours, such as a window-fastener handle are described. The machine incorporates four rams in line, the sequence of operation of which is arranged so as to cause the metal to flow in the desired manner.—R. A. R.

Iron Powder Metallurgy with Special Reference to Sintered Steel and Its Application. H. Bernstorff and H. Silbereisen. (Archiv für Metallkunde, 1948, vol. 2, No. 9, pp. 295–300). The properties of parts made from iron and from steel powders by powder metallurgy are reviewed. The economic aspects of production by machining, forging, stamping, and powder metallurgy are compared, and the limitations of powder metallurgy are pointed out.—R. A. R.

PROPERTIES AND TESTS

General Solution of the Problem of Plane Plasticity for any Isotropic or Anisotropic Law of Flow. H. Neuber. (Zeitschrift für angewandte Mathematik und Mechanik, 1948, vol. 28, Sept., pp. 253–257 : Physics Abstracts, 1949, vol. 52, Feb., p. 88). The solutions of classical theory are only valid for isotropic materials where Mohr's limiting curve is a straight line. By means of a "surface of plasticity" (*i.e.*, three-dimensional representation of the plane plastic flow in a co-ordinate system formed by the three stress components) the general solution of any law of plastic flow may be given by a simple graphical method. The characteristics curves of the equations of equilibrium are the curves of 45° gradient on the surface of plasticity.

Conditions of the Occurrence of Plastic Deformation in Solids of Simplest Shape Quenched from the Surface. B. N. Finkelstein. (Journal of Technical Physics, U.S.S.R., 1948, vol. 18, Aug., pp. 1026–1028).

Fracture Characteristics of Ship Plate in Certain Small-Scale Tests. E. P. Klier, F. C. Wagner, and M. Jensamer. (Welding Journal, 1949, vol. 28, Feb., pp. 50-s–66-s). The authors have modified the slow-bend notch-bar test (*see* Journ. I. and S.I., 1948, vol. 159, July, p. 337), by introducing a drilled hole through the compression zone which is filled with a hardened pin to which the load is applied. Notch-bar tensile tests were also carried out on specimens 2·75 in. wide with sawcuts 0·031 in. wide leaving the portion in tension 0·625 in. wide. They find the slow-bend notch-bar test gives satisfactory data for the change from ductile to brittle fracture with falling temperature, and correlation with the results obtained with the 72-in. wide internally notched plate used in earlier work. The edge-notched tensile tests, however, give two transition ranges, one associated with change of type of fracture, and the other with a fall in the energy absorption : the former agrees approximately with the larger-scale tests, the latter does not.—J. P. S.

Axial Tension Impact Tests of Structural Steels. W. H. Bruckner and N. M. Newmark. (Welding Journal, 1949, vol. 28, Feb., pp. 67-s–88-s). Attempts using Charpy V-notch and keyhole-notch flexural impact tests to obtain correlation with wide-plate tests have not been satisfactory (*see* preceding abstract). Correlation was obtained with rimmed and killed steels, but two semi-killed steels exhibiting large differences in wide-plate tests showed practically no difference in Charpy tests : in the latter tests, moreover, the transition temperature (from ductile to brittle fracture) was difficult to define. Notched-bar impact tests, on a cylindrical specimen of outside diameter 0·625 in., with a notch 0·01655 in. deep, having a root radius of 0·039 in. giving a root diameter to the specimen of 0·294 in., were carried out at a number of values of initial energy of the pendulum of the testing machine. Transition temperatures were determined from both energy absorption and reduction in area ; the results were similar. Where the initial energy was 110 ft.lb. or less, transitions were sharper : lower initial energies did not affect the transition temperatures of the rimmed steel and one of the semi-killed steels under investigation, but lowered the transition temperatures of the other semi-killed steel and a killed steel. At 110 ft.lb. or 45 ft.lb. initial energy the transition temperatures of the semi-killed steels differed by 30° to 40° F., and all the transition temperatures agreed fairly well with the findings of the wide-plate tests.—J. P. S.

The N : S Relationship in Endurance Testing. M. G. Corson. (Iron Age, 1949, vol. 163, Mar. 10, pp. 103–105). The author has derived a mathematical formula which relates the number of cycles of stress to failure (N) to the stress (S) and includes a factor allowing for the lag in adjustment of the test-piece to plastic strain between reversals of stress. Application of this to published data confirms its usefulness.—J. P. S.

Influence of the Shape and Dimensions of the Test Piece on the Fatigue Limit. P. Laurent. (Société Française de Métallurgie, Oct. 10, 1947 : Revue de Métallurgie, Mémoires, 1949, vol. 46, Jan., pp. 55–59). The mechanism of the development of fatigue under alternating stresses and of failure by fatigue, and the influence of the shape and dimensions on the fatigue limit are discussed ; reference is made to the work of numerous investigators.—A. E. C.

Machine for Fatigue Testing at High Temperatures. M. P. Markovets, I. A. Smiyan, and N. I. Mikheev. (Zavodskaya Laboratoriya, 1949, vol. 15, Jan., pp. 82–85). [In Russian]. A rotating cantilever fatigue-testing machine is described for use at temperatures up to 900° C. The temperature of the test is automatically regulated to ± 2° C., and the maximum stress that can be applied to the specimen is 60 kg./sq. mm. The importance of correctly measuring the surface temperature of the specimen in fatigue tests is pointed out, and a special device for this is incorporated in the design of the machine. Specimens up to 10 mm. in dia. can be dealt with. The overall dimensions of the machine are : Width, 500 mm. ; length, 1000 mm. ; height, 1100 mm.—S. K.

Machine for Testing a Fixed Specimen for Fatigue at High Temperatures. S. I. Yatskevich. (Zavodskaya Laboratoriya, 1949, vol. 15, Jan., pp. 86–88). [In Russian]. An account is given of a fatigue-testing machine in which the specimen, 8 mm. in dia., is held vertically inside an electric muffle and subjected to cyclic bending stresses at 50 cycles/sec., the maximum stress in the specimen being 50 kg./sq. mm. The specimen is gripped firmly at its upper end in a clamp specially designed to obviate the necessity of accurately machining this part, and a rotating semi-flexible arm imparts the cyclic stress to the lower end through a suitable system of ball-bearings and gears. The temperature of the furnace is controlled automatically. The machine is stated to be specially suitable for testing heavy parts such as the axles of railway rolling-stock.—S. K.

Machine for Fatigue Testing during Plane Bending. A. N. Mitinskii and V. A. Bykov. (Zavodskaya Laboratoriya, 1949, vol. 15, Jan., pp. 89–91). [In Russian]. In the testing machine described, the specimen can be subjected to bending in a single plane at four different frequencies, ranging from 16 to 500/min., and with variable amounts of deformation. Among the distinctive features of the machine are the following : Simplicity of the mechanical system deforming the specimen ; changing sign of the bending of the specimen in one plane ; small frequency of the cycles ; the possibility of testing comparatively large specimens without changing their surface conditions ; possibility of testing with either symmetric or other cycles.—S. K.

On the Influence of Short Periods of Overloading on the Endurance Limit of Steel. G. I. Pogodin-Alekseev. (Zavodskaya Laboratoriya, 1948, vol. 15, Jan., pp. 91–95). [In Russian]. After a brief critical review of previous work on the influence of overloading of short duration on the endurance limit of steel, an account is given of experiments carried out on cylindrical specimens, 10 mm. in dia., of axle steel. In these experiments the specimens were subjected to overloading to the extent of 31·6% for a given fraction of the previously determined mean number of cycles required to produce failure, the number of cycles needed to produce failure being determined. It was found that, when the above fraction corresponded to 25%, the endurance limit was raised by 4–5%, whilst a fraction corresponding to 60% or more led to lowered endurance limits. These results are discussed and their importance is pointed out.—S. K.

Testing Machine IM-4R and Its Defects. M. A. Shapiro and A. S. Krichever. (Zavodskaya Laboratoriya, 1949, vol. 15, Jan., pp. 100–101). [In Russian]. Defects in the design and construction of a Russian testing machine are pointed out. S. K.

Application of the Pulsator to the Testing of Heavily Loaded Toothed Gears of High Modulus. R. S. Nikolaev, E. F. Mikhnenko, and L. M. Shkolnik. (Zavodskaya Laboratoriya, 1949, vol. 15, Jan., pp. 124–125). [In Russian]. The application of the pulsator in a machine for testing the fatigue strength of heavy toothed gears is described.—S. K.

Fatigue Limit of S.A.E. 1095 after Various Heat Treatments. A. C. Forsyth and R. P. Carreker. (Metal Progress, 1948, vol. 54, Nov., pp. 683–685). SAE 1095, a plain hypereutectoid steel, as cast, as rolled, or as annealed, has a relatively low fatigue limit : 30,000 to 50,000 lb./sq. in. In the experiments described specimens of SAE 1095 steel were heat-treated in three ways : By water-quenching and then tempering ; by austempering ; and by martempering and then tempering ; a hardness of Rockwell C53 was obtained in each case. The fatigue strength of the martempered specimen was 160,000 lb./sq. in. which was considerably higher than the water quenched and tempered treatment figure 124,000 lb./sq. in. and the austempered treatment figure 130,000 lb./

sq. in. The method of testing is described and data obtained are presented. It is suggested that, although more expensive, martempering heat-treatment may prove economically advantageous when the fatigue failure of steel parts with high hardness is a problem.—J. C. R.

Effects of Inclusions on the Endurance Properties of Steels. W. C. Stewart and W. L. Williams. (Journal of the American Society of Naval Engineers, 1948, vol. 60, Nov., pp. 475–504). Inspectors of steel for large forgings have difficulty in determining when the number and size of non-metallic inclusions warrants rejection of the piece. This paper describes tests at the U.S. Naval Engineering Experiment Station, Annapolis, in which specimens containing inclusions were subjected to rotating cantilever and alternating torsion fatigue tests. Theoretical considerations are also examined, and the general conclusions are : (1) The inclusions which are most serious in machinery components are usually those exposed on the surface where stresses are highest ; (2) Severe inclusions in the surface can lower the fatigue strength considerably ; (3) when a surface is not uniformly stressed the position of an inclusion may be the deciding factor, *e.g.*, an inclusion in a fillet, keyway, or oil hole would be more undesirable than in plain surfaces ; (4) longitudinal inclusions parallel to the direction of stress do not significantly reduce the flexural fatigue strength, but large ones can markedly reduce the torsional fatigue strength ; (5) a number of longitudinal inclusions forming a non-longitudinal pattern can be much more damaging than similar inclusions randomly distributed ; (6) small globular inclusions apparently have little effect ; and (7) no clear specification can be written to describe the acceptable limits of non-metallic inclusions in large forged parts.—R. A. R.

On the Theory of Hardening by Plastic Deformation. G. Masing. (Zeitschrift für Physik, 1948, vol. 124, No. 7–12, pp. 586–601 : Physics Abstracts, 1949, vol. 52, Feb., p. 89). This is a critical review. Hardening is due to increasing difficulty in the movement of dislocations, not in their production. Taylor's theory of the mutual action of dislocations is insufficient, as it predicts a considerable dependence of the rigidity on the degree of deformation. A modification of Mott's theory of the interaction of dislocations and internal stresses explains the phenomena qualitatively.

The Interpretation of the Jominy End Quench Hardenability Test. L. C. Bogan. (Australian Institute of Metals : Australasian Engineer, 1948, Dec. 7, pp. 47–59). The literature of the last decade on the Jominy end-quench test is critically reviewed.—R. A. R.

The Microhardness of Carbides in Tool Steels. L. P. Tarasov. (Metal Progress, 1948, vol. 54, Dec., pp. 846–847). The author has endeavoured to clarify the marked resistance to grinding of some high-alloy types of tool steel by measuring the Knoop hardness of the carbide particles in three steels : a plain carbon tool steel containing 1·3% of carbon, which was annealed to make the cementite massive enough to be indented ; a high-carbon high-chromium steel (C 2·2%, Cr 12·5%, V 1%) which was hardened and tempered ; and a high-carbon high-vanadium high-speed steel (C 1·5%, W 12–14%, Cr 5%, V 5%, Co 5%) which was hardened and tempered. The data are given in tabular form together with data from tests on aluminium oxide, complex vanadium carbides, and silicon carbide. The author discusses the results obtained.—J. C. R.

Note on the Determination of Surface Hardness. A. Portevin. (Métaux et Corrosion, 1947, vol. 22, Feb., pp. 31–33). A brief survey is presented of methods of hardness determination, subdivided according to the principle or definition of hardness, the conditions of deformation, the depth of metal affected by permanent deformation, and the method of assessing the results obtained.—J. C. R.

Brittle Fracture in Ship Plate. G. Vedeler. (Engineering, 1949, vol. 167, Apr. 1, pp. 289–290). The conditions of stress leading to brittle fracture in mild steel plate are considered. It is shown that brittle fracture can occur only when all the principal stresses are large and the largest of them is much higher than the breaking strength, and that reducing the temperature by 50° C. is equivalent to an increase of strain rate of at least a thousandfold. Changing the load from static to dynamic increases both the yield point and the ultimate strength—the former much more than the latter. A stress concentration has an effect similar to that of lowering the temperature.—R. A. R.

Elementary Friction. Wear. A. Marcelin. (Société Française de Métallurgie, Oct. 10, 1947 : Revue de Métallurgie, Mémoires, 1949, vol. 46, Jan., pp. 27–35). The author first indicates the three modes of lubricated friction (hydrodynamic, minimum liquid film, stratofilm) and briefly describes the conditions prevailing in each. He points out that contact between the two surfaces rubbing against one another occurs at a number of points ; as the load is raised they become minute surfaces, and their number increases. The author calls " elementary friction " the friction at one point of contact. After touching on the " frictiograph " used in his experiments the author presents the results of three series of tests. In the first an ordinary bronze milling tool was pressed against a medium-hard steel reciprocating block, with a given lubricant and this " triplet " was studied in detail ; in the second series different lubricants were used, and in the third series the materials (cast iron, bronze, steel, aluminium alloy) used in making the milling tool were varied. A. E. C.

Influence of Electrolytic Polishing on the Phenomena of Friction and Wear. R. Mondon. (Société Française de Métallurgie, Oct. 10, 1947 : Revue de Métallurgie, Mémoires, 1948, vol. 45, Dec., pp. 521–524). The object of the experiments was to determine the influence of the surface modification due to electrolytic polishing on the friction and wear of steels used for motor parts. The first tests were made on a reciprocating " frictiograph " (for description *see* preceding abstract of a paper by A. Marcelin) with a special bronze rubbing wheel bearing on lubricated blocks of a case-hardened nickel–chromium–molybdenum steel, a medium-hard nickel–chromium–molybdenum steel in two conditions of heat-treatment, and a nitrided chromium–molybdenum steel ; all specimens were finished by fine grinding and some were subsequently polished electrolytically. The friction on starting, the friction after running for 3 min., and the wear on the wheel after 3 min. operation were measured. The second series of tests was made in a machine in which a shaft rotated in a collar, under lubrication, and the frictional couple at starting and at stopping was measured. Tests were made on the first three steels mentioned above. In both series electrolytic polishing was beneficial in regard to both friction and wear, but the authors cannot say whether this is due to the grinding of the micro-profile or to a modification of the surface layers. Under the test conditions in the first series the highest surface hardness did not necessarily lead to the least friction and least wear.

Determination of the Sliding Coefficient of Friction. I. A. Savostin. (Zavodskaya Laboratoriya, 1949, vol. 15, Jan., pp. 96–99). [In Russian]. An account is given of an apparatus in which the sliding coefficient of friction of various pairs of surfaces can be measured in the dry or wet state, at high or low temperature, for various sliding speeds and under high or low load. One of the surfaces is in the form of a flat ring which is rotated at 3·1 to 155·5 r.p.m., the other surface being held in contact with the ring and the frictional force between the two being determined with the aid of a torsional mirror dynamometer. The maximum load which can be applied to the stationary surface is 10 kg., and provision is made for supply of liquid, temperature control, etc. The results obtained for a number of materials, including steels, under various conditions are presented.—S. K.

The Frictional Behaviour of Engineering Materials. R. C. Parker. (Engineering, 1949, vol. 167, Mar. 4, pp. 193–196 ; Mar. 11, pp. 217–219). An account of the main developments of friction research is given which shows that the engineer has constantly to meet conditions for which there is no law. Most of the test data presented refer to moulded and metallic and non-metallic material for woven brake-lining materials. The large inertia machine in the Research Laboratories of Ferodo, Ltd., is described.—R. A. R.

Deltamax—A New Magnetic Core Material. W. S. Spring. (Iron Age, 1949, vol. 163, Mar. 31, pp. 70–73). Deltamax is a new 50% nickel–iron alloy rolled and heat-treated so that the most easily magnetized axes of the crystals are parallel to the direction of rolling. It resembles the German Permenorm 5000 Z in composition and in having a nearly rectangular hysteresis loop, but the process of manufacture has been simplified by dispensing with vacuum melting and by postponing slitting until the final thickness is reached.—J. P. S.

A Review of Magnetic Materials Especially for Communication Systems. R. A. Chegwidden. (Metal Progress, 1948, vol.

54, Nov., pp. 704B–714B). In this review developments in the knowledge of the properties of high permeability materials and alloys used for permanent magnets are discussed. A table is given in which these properties are summarized. A bibliography of 19 references is appended.—J. C. R.

On the Theory of the Magnetization of an Elastically Distorted 25% Cr–Fe Alloy. O. Rüdiger. (Zeitschrift für Physik, 1948, vol. 124, No. 7–12, pp. 458–468: Physics Abstracts, 1949, vol. 52, Feb., p. 78). The influence of mechanical strains on the spin distribution in a 25% chromium–iron alloy, in which only one magnetoelastic constant differs from zero, is considered. The spins set themselves always along the cube edges.

Detection of Ferrite by Its Magnetism. T. V. Simpkinson and M. J. Lavigne. (Metal Progress, 1949, vol. 55, Feb., pp. 164–167). The use of the Magne-gage for the detection of ferrite in stainless steels is described. This instrument can be used to measure the relative ferromagnetism of slightly magnetic specimens. As fully austenitic stainless steel, the sigma phase, and carbides in stainless steel are not ferromagnetic, any ferromagnetism detected must be due to the presence of ferrite. Phase changes involving ferrite, brought about by heat-treatment, may therefore be followed by measuring the ferromagnetism of samples with this instrument.—J. C. R.

The Influence of Direct Electric Current on the Diffusion of Substances Precipitated in Metals. G. Bianchi and R. Zoja. (Ricera Scientifica e Ricostruzione, 1947, vol. 17, Dec., pp. 1987–1990: Physics Abstracts, 1949, vol. 52, Feb., p. 74). Tests were made on bars of four kinds of steel (two containing precipitated oxides). Direct currents, up to 900 amp. at temperatures up to 450° C. were passed through them for periods up to 480 hr. The bars were tested before and after each experiment for electrical resistance and elasticity, or the microscopic structure was examined. No changes which could be attributed to ionic diffusion were observed. Control experiments were made with A.C.

Coordination of Metallurgical Work at General Electric. W. E. Ruder. (Metal Progress, 1949, vol. 55, Jan., pp. 43–46). A review is presented of the work done and in progress in the different research laboratories of the General Electric Company of America.—J. C. R.

A Proposal for Research in Metallurgy. M. G. Corson. (Metal Progress, 1949, vol. 55, Jan., pp. 55–58). The author outlines a programme of research on various metallurgical topics and the estimated cost, where it is unlikely that this work would be undertaken by industrial companies. The work would include preparation of purified metals, compressibility, specific gravity and thermal expansion, electrical and thermal conductivity, magnetic susceptibility, photo-electric effect, heat content, electrochemical and contact potentials, and mono-directional and bidirectional stress.—J. C. R.

Ultrasonic Thickness Indicator. B. Carlin. (Electronics, 1948, vol. 21, Nov., pp. 76–79: [Abstract] Bulletin and Foundry Abstracts of the British Cast Iron Research Association, 1949, vol. 9, Jan., p. 356). This is a description of the Reflectogage which is based on the principles used in the Sonigage. The instrument measures the thickness of material forming pipes, tanks, and vessels which can only be reached from one side of the wall. Ultrasonic waves are generated, transmitted through the article under test, and reflected within the material either by its opposite side, or a flaw or discontinuity. The commercial instrument reads directly in the range from 0·025 to 0·300 in. in steel, and up to 4 in. by harmonic methods. It can also be used for plastics and glass. Details are given of electronic and ultrasonic principles upon which the design is based, and of the circuit. Readings are provided by visual display upon a graduated screen.

Non-Destructive Testing of Drill Pipe. R. C. McMaster. (Non-Destructive Testing, 1948, vol. 7, Autumn Issue, pp. 27–33). Methods of testing strings of drilling pipe to determine when renewals are required are discussed; special reference is made to Magnaflux testing. There are 26 references.—R.A.R.

Non-Destructive Testing and Its Place in the Current and Prospective Navy Programs. R. H. Lambert. (Non-Destructive Testing, 1948, vol. 2, Autumn Issue, pp. 20–21). The use of non-destructive testing, particularly X-rays, by the U.S. Navy is discussed in general terms.—R. A. R.

A Simple Constant-Stress Creep Test. J. C. Fisher and R. P. Carreker. (American Institute of Mining and Metallurgical Engineers, Technical Note No. 10: Journal of Metals, 1949, vol. 1, Feb., Section 3, p. 178). A method is described for obtaining constant stress which is especially suited to the testing of small wires; it can also be adapted to testing rods of any diameter.—J. C. R.

A Study of Residual Gases in Cast Iron. J. E. Hurst and R. V. Riley. (Proceedings of the Institute of British Foundrymen, 1947–48, vol. 51, pp. A55–A66). *See* Journ. I. and S.I., 1949, vol. 161, Feb., p. 172.

Gases and Metals. H. Lepp. (Vide, 1948, May, pp. 433–441: [Abstract] Bulletin and Foundry Abstracts of the British Cast Iron Research Association, 1948, vol. 9, Nov., p. 307). The author discusses the phenomena of adsorption and absorption of gas by metals. He considers thermodynamic analysis to be the most efficient method of solving problems arising from the use of vacuum technique in metallurgy.

Contribution to the Problem of the Pickling Embrittlement of Single Iron Crystals. G. Chaudron and L. Moreau. (Archiv für Metallkunde, 1948, vol. 2, No. 9, pp. 308–311.) The effect of treating monocrystalline and polycrystalline specimens of pure iron with hydrogen by pickling in hydrochloric acid and by electrolysis was studied by making hardness and bending-fatigue tests before and after the treatment. In the annealed state a single crystal was not embrittled by pickling, but a cold-reduced single crystal (reduced from 0·8 to 0·3 mm. in thickness) showed a marked increase in brittleness. Cold-reducing polycrystalline iron accelerated the embrittlement induced by pickling and electrolysis. R. A. R.

Influence of Elements on the Characteristics of Steels. J. V. Lopez. (Anales de Mecanica y Electricidad, 1948, vol. 24, pp. 146–149: British Abstracts, B.I, 1948, Dec., col. 716). The effects of additions of beryllium and boron on the characteristics of steels of different types are reviewed.

The Specification and Testing of Cast Iron. A. B. Everest. (Proceedings of the Institute of British Foundrymen, 1947–48, vol. 51, pp. B75–B81). *See* Journ. I. and S.I., 1948, vol. 159, July, p. 343.

Ductile Cast Iron. A. P. Gagnebin, K. D. Millis, and N. P. Billing. (Iron Age, 1949, vol. 163, Feb. 17, pp. 77–84). The addition of magnesium or a nickel–magnesium alloy to pearlitic iron containing carbon 3·2–3·6%, and silicon 1·8–2·8% converts the graphite into a spheroidal form. The material as cast possesses a 0·2% proof stress of 60,000 lb./sq. in. (26·7 tons/sq. in.), a tensile strength of 96,000 lb./sq. in. (42·8 tons/sq. in.), an elongation of 7% and a Brinell hardness of 220; its proof stress, tensile strength and hardness are reduced to two-thirds of these values by annealing, but its elongation is raised to 18–20%. Normalizing at 1650° F. (900° C.) followed by tempering at 1100° F. (590° C.) raises the proof stress to 116,000 lb./sq. in (51·7 tons/sq. in.), the tensile strength to 127,000 lb./sq. in. (56·5 tons/sq. in.) and the hardness to 304 Brinell; the elongation is reduced to 2·0%. The resistance to growth at 1600° F. (870° C.) is five times greater than that of a grey cast iron. Excellent machining properties, weldability, casting properties, and high endurance values are also claimed.—J. P. S.

Alloy Iron Castings Show Tensile Strength almost 60,000 Psi. as Cast. (Inco Magazine, 1948, vol. 22, Summer Issue, pp. 4–6: [Abstract] Bulletin and Foundry Abstracts of the British Cast Iron Research Association, 1949, vol. 9, Jan., p. 348). Properties and applications of Ni–Tensyl irons, grey iron alloys containing up to 3% of nickel and 0·6% of molybdenum, are given.

Central Inspection Department Solves Daily Production Problems at Hunter Spring. (Steel, 1949, vol. 124, Feb. 21, pp. 123–124). Inspection of springs at the works of Hunter Spring Co., Lansdale, Pa., is carried out in a centrally located laboratory. Unspecified tests are carried out on the first few items of a run of manufacture, at 15–20 min. intervals during the run, and in some cases on 100% of the production. A report on "quality control," based on sampling and frequency distribution tests is available to buyers.—J. P. S.

Recent Advances in Alloy and Special Steels. D. A. Oliver. (Metallurgia, 1948, vol. 39, Dec., pp. 71–74). Advances in the fields of steels for general use at elevated temperatures, for gas-turbines, tool steels, and permanent magnets, are reviewed.

Some Developments in Alloys Containing Nickel. L. B. Pfeil. (Metallurgia, 1948, vol. 39, Dec., pp. 81–86). Some of

the less widely known advances in the development of nickel alloys are reviewed. These include: (1) A shock-resisting alloy cast iron (Ni 1·0–1·5%, Mo 0·3–0·5%); (2) an acicular cast iron (Ni 1–3%, Mo about 1%) which, after a simple tempering treatment, has a tensile strength of the order of 40 tons/sq. in.; (3) Ni-hard white alloy cast iron for shot-blast nozzles, mixer knives, and crushing machinery; (4) Alcomax and Ticonal permanent-magnet alloys; (5) Kovar and Nilo K low-expansion alloys; (6) alloys with constant elastic properties, such as Invar; (7) stainless steels with improved hot-working properties (Cr 16–18%, Ni 9–12%, Mo 1·5–3%); and (8) free-machining stainless steels.—R. A. R.

Alloy Steel in Oil-Field Equipment. R. L. Adams. (Metal Progress, 1948, vol. 54, Oct., pp. 468–470). A survey is presented of the use of alloy steels for drills, pumping machinery parts, the lining of the well hole, and other auxiliary equipment.—J. C. R.

Alloy Steel in Oil Refineries. F. C. Braun. (Metal Progress, 1948, vol. 54, Oct., pp. 471–473). The author outlines the use of alloy steels for tubes, bolts, valves, and other parts of oil-refinery machinery. At least half the alloy steel is of the 5% chromium–molybdenum or 7% chromium–molybdenum type, and about one-tenth is 9% chromium–molybdenum. The rest is distributed among the chromium–molybdenum steels of lower alloy content and 18/8 stainless steel, which is used for the more severe conditions.—J. C. R.

Alloy Steel in the Turret Lathe. D. M. Gurney. (Metal Progress, 1948, vol. 54, Oct., pp. 474–476). The author reviews the gradual replacement of all the cast iron and unalloyed steel parts of a turret lathe by alloy steel which has been oil-quenched and tempered.—J. C. R.

Low-Alloy Steels in the Electrical Industry. J. T. Rettaliata. (Metal Progress, 1948, vol. 54, Oct., pp. 477–481). A review is given of the use of low alloy steels (notably silicon–iron alloys) for transformers, generators, motors, and other equipment used in the electrical industry.—J. C. R.

Alloy Steel in War—Then and Now. C. Williams. (Metal Progress, 1948, vol. 54, Oct., pp. 485–488, 490). Notes are given on the ever-increasing use of alloy steels for armanent and ordnance purposes over the last 75 years.—J. C. R.

Structural Alloy Steels in the Air Age—Aircraft Engines. A. W. F. Green. (Metal Progress, 1948, vol. 54, Oct., pp. 491–495). Developments in the use of alloy steels for aircraft engines are discussed.—J. C. R.

Alloy Steels in the Airplane Itself. L. D. Bonham. (Metal Progress, 1948, vol. 54, Oct., pp. 495–496). A brief note is given on the increased use of alloy steels for aeroplane parts. Low-alloy steels can be used in structural fittings, landing gear, and other equipment where high strength is required, but it is desired to save weight, and light alloys would be unsatisfactory.—J. C. R.

Rôle of Alloy Steels in the Automotive Industry. W. F. Sherman. (Metal Progress, 1948, vol. 54, Oct., pp. 497–501). An account is given of the development of the motor-car industry in America with notes on its increased use of alloy steels.—J. C. R.

Alloy Steels in Railroad Service. J. L. Carver. (Metal Progress, 1948, vol. 54, Oct., pp. 502–506). The author discusses the extent to which alloy steels can be used for railway engine parts and structural parts of railway trucks. J. C. R.

Alloy Steels, the Farm Tractor, and the Full Granary. M. L. Frey and G. C. Riegel. (Metal Progress, 1948, vol. 54, Oct., pp. 507–510). The authors review the use of alloy steels in the manufacture of farm machinery and discuss the heat-treatment of these steels.—J. C. R.

Improved Silicon Irons for Electrical Equipment. W. Morrill. (Metal Progress, 1948, vol. 54, Nov., pp. 675–678). The early history of silicon–iron alloys is reviewed and the author then gives an account of investigations undertaken since 1933 which have enabled the degree of preferred orientation to be increased to 95%. When the nitrogen content was increased from 0·0005% to 0·0010% the number of secondary crystals which grew at the expense of the recrystallized matrix became fewer, their size became larger and their orientation more perfect. However, when the nitrogen content approached 0·0015% the growth-inhibiting effect prevented the growth of crystals. Some notes are given on the economic advantages of high quality silicon irons.—J. C. R.

Gray Iron Transverse Test Bars. J. H. Schaum. (Foundry, 1948, vol. 76, Dec., pp. 68–71, 216; 1949, vol. 77, Jan., pp. 82–85, 222). Common defects found in commercial transverse test-bars and technique for improving the casting and testing of these bars are described. The effect of the size of the specimen and its location in the casting on its tensile strength is studied. Data derived from tests on 29 standard test-bars are given and the following conclusions are reached: (1) Test-bars are poured in a variety of manners some of which lead to elliptical cross-sections and surface defects; (2) tensile properties of cast iron may vary considerably from one end of a test bar to the other; the strongest metal is located near the bottom of the vertically cast top-poured bars; (3) the relation between transverse and tensile strength of single test-bars is so inexact as to preclude any accurate prediction of either of these properties when the other is known; (4) from the appearance of fractures it is possible to estimate the approximate tensile and transverse strength and equivalent carbon of test-bars if diameter and pouring temperature are held constant; (5) best qualities are obtained by (*a*) using a chromium-plated steel pattern, (*b*) using a fine-grained sand, (*c*) ramming the mould in the horizontal position, (*d*) pouring the bar in a horizontal position and setting it vertical as soon as possible; (6) the smaller the tensile specimen taken from a casting of given cross-section, the higher is the maximum and the lower the minimum observed strength. A rapid, accurate, and inexpensive transverse testing technique is described which consists of a dial, calibrated in units of 0·001 in., anchored to the testing-machine ram with the dial pin in contact with a rigid reference bar.—J. C. R.

Meeting High Performance Demands in Engineering Construction with Meehanite. A. Campion. (Engineering Centre, Glasgow: Engineer, 1949, vol. 187, Mar. 4, pp. 243–245: Metallurgia, 1949, vol. 39, Mar., pp. 272–276: Machinery, 1949, vol. 74, Mar. 31, pp. 405–409; Apr. 7, pp. 440–442.) **High Duty Cast Iron.** A. Campion. (Machine Shop Magazine, 1949, vol. 10, Mar., pp. 104–108, 128–132). **Engineering Construction.** A. Campion. (Iron and Steel, 1949, vol. 22, Apr., pp. 131–135; May, pp. 154–156). **The Engineering Properties of Meehanite.** A. Campion. (Foundry Trade Journal, 1949, vol. 86, Apr. 14, pp. 335–338; Apr. 21, pp. 369–372, 375). The mechanical properties of Meehanite are considered and many test data are presented by which its properties can be compared with those of cast iron and steel.—R. A. R.

The Effect of Alloys on the Properties of Steel. T. W. Merrill. (Product Engineering, 1949, vol. 20, Mar., pp. 124–127). The manner in which molybdenum, vanadium, copper, nickel, chromium, and other alloying elements affect the hardenability and mechanical properties is discussed. R. A. R.

Noteworthy Developments in the Sphere of Tool Steels. H. Krainer and J. Frehser. (Betrieb und Fertigung, 1949, vol. 3, Apr., pp. 49–55). Developments in tool steels in the countries of Central Europe since 1936, with special reference to a reduction in the consumption of alloying elements, are reviewed. By combinations of up to five alloying elements it has been possible to reduce considerably the total content of these elements required in high-speed steels and hot and cold working steels.—R. A. R.

Structural High-Tensile (Low Alloy) Steel. O. A. Kerensky. (Third Congress of International Association for Bridge and Structural Engineering: Engineer, 1949, vol. 187, Mar. 4, pp. 238–241). The mechanical properties, weldability, and corrosion-resistance of structural high-tensile steels are discussed. Data on the steels used for a number of British-built bridges and the stresses to which members are subjected are presented, and comparative costs of high-tensile and mild steels are considered.—R. A. R.

The Properties of Castings with 12% Mn (Hadfield's Cast Steel). F. Roll. (Archiv für Metallkunde, 1949, vol. 3, Jan., pp. 18–23). The properties of 12% manganese cast steel are reported with special reference to the effect of the carbon/manganese ratio, and the silicon and phosphorus contents. Brinell and microhardness data are evaluated and the microstructure is described.—R. A. R.

About 12% Manganese Cast Steel. H. Resow. (Neue Giesserei, 1949, vol. 36, Mar., pp. 67–74). The toughness and work-hardening properties of cast 12% manganese steel are discussed. The hardening is caused by a definite form and arrangement of a carbide precipitated from the austenite.

The effect of the carbon content, alloying additions, and quenching temperature on the toughness and hardness of cold-deformed specimens was investigated. Tensile, impact, and compressive tests were also made to ascertain which kind of deformation produced the greatest hardness. Finally, the relation of hardness to wall thickness and the effect of this in practical applications are dealt with.—R. A. R.

Guidance on the Application of Unalloyed Steels Made by Different Processes. K. Daeves and H. Hauttmann. (Verein deutscher Eisenhüttenleute, 1943, Confidential Report No. 44). The purpose of the paper is to present in summary and tabulated form a survey of the properties of unalloyed steel made by different processes. In general, the definitions, properties, applications, and tables show that unalloyed open-hearth steel is necessary in practice from the point of view of quality for a minimum tensile strength of over 43 tons/sq. in. and for specially high purity, just as it is for highly stressed case-hardened and heat-treated parts. In all other cases, the use of Bessemer steel is possible without impairing the purpose of the application, indeed, normal basic-Bessemer steels, taking note of their special properties and with intelligent working, design, and application, are available for a very wide range of applications.—G. W. A.

Material Economics and Improvement in Strength by Heat-Treating Basic-Bessemer Steel. F. Kösters. (Verein deutscher Eisenhüttenleute, 1943, Confidential Report No. 52). It is established that by the heat-treatment of basic-Bessemer steels: (1) High yield points can be obtained whereby many high-quality steels with yield points of up to 80 kg./sq. mm. can be replaced by heat-treated basic-Bessemer steel; (2) the notched-bar impact strength of basic-Bessemer steel in the imaged condition, tested at + 20° C. is improved; (3) the brittleness produced by ageing in low-carbon steels, already apparent in basic-Bessemer and open-hearth steels above 50° C. is shifted to below 0° C. by heat-treatment, and even lower in higher-carbon basic-Bessemer steels; (4) thin rolled sections, despite ageing, show extraordinarily high notched-bar impact results when tested at 50° C.; (5) basic-Bessemer steels, deoxidized with aluminium and heat-treated, show an appreciable improvement through heat-treating (apparent at testing temperatures from − 50° C. upwards) over and above the improvement in notched-bar results after the aluminium deoxidation. Considerable savings in steel are attainable by raising the permissible working stress consequent on the higher yield point: there is also an improvement in the notched-bar impact results. As a consequence of a quench in warm water for the heat-treatment, the operations are simple; the temperature after rolling can be employed, and large quantities of hot water are made available for use in the works.—G. W. A.

The Austenitic Stainless Steels—American and British Practice Compared. F. H. Keating. (Metal Progress, 1948, vol. 54, Dec., pp. 819–822). The author examines the differences between American and British practice in the production of chromium–nickel austenitic steels. In the United States only 5% of these steels are made of stabilized compositions (compositions immune to intercrystalline disintegration) as compared with a very much higher percentage in Britain. Secondly, in America the carbon content is kept much lower (normally 0·06–0·08% but even as low as 0·03%) than in Britain (up to 0·16%). Thirdly, while both countries use special steels containing molybdenum, a significant proportion of British production is stabilized, but little of the American production contains a stabilizing agent and a final heat-treatment is used. Fourthly, niobium is used in American and titanium in Britain as a stabilizing element. Reference is made to the Huey test in America and the Strauss test in Britain to assess the performance of these steels.—J. C. R.

Effect of Strength and Ductility on Burst Characteristics of Rotating Disks. A. G. Holms and J. E. Jenkins. (National Advisory Committee for Aeronautics, 1948, July, Technical Note No. 1667, pp. 1–52: [Abstract] Applied Mechanics Reviews, 1948, vol. 1, Sept., pp. 235–236). This paper records the results of tests to determine bursting speeds of rotating discs made of various materials having widely varying strength and ductility properties. The discs of 10 in. outside dia. and of approximately $\frac{3}{8}$ in. thickness, were driven by an air turbine. Tests were made both on solid discs and on discs with $1\frac{1}{2}$ in. and $\frac{1}{16}$ in. dia. holes. Materials tested included SAE 1078 and SAE 1035 steels, stainless steel, beryllium, copper, brass, and aluminium- and nickel-base alloys. Ductility in the materials used varied as indicated by a range of tensile elongation values from 3·4% to 52·8%. In most cases failures were of the brittle or cleavage type, at 90° to the plane of the disc, with the latter splitting into a number of fragments. The most ductile discs, however, failed at angles of 45° to the plane of the disc. Some discs with intermediate values of ductility showed both types of failure. In general, it was found that for both solid discs and discs with holes, the disc strength (in terms of speed necessary to cause fracture) increased with tensile strength regardless of ductility. Introduction of a large hole reduced the strength approximately in proportion to the amount of material removed. A slight reduction in strength was also obtained for discs with $\frac{1}{16}$-in. holes, but nowhere near the 50% reduction called for on the basis of the elastic-stress theory. In the case of solid discs, the ratio of the calculated elastic stress at failure to tensile strength ranged between the value of 1·0 predicted by the elastic-stress theory and the value of 1·24 predicted by the average-stress theory. For discs with holes the values of this ratio were, in general, somewhat below those predicted on the basis of the average-stress theory, but considerably above those predicted from the elastic-stress theory. It should be noted here that neither of these theories takes into account stress increases due to expansion in disc diameter or reduction of disc thickness at fracture. Tests on discs made of a very hard material (Rockwell C 60) of low ductility showed a reduction in strength around 30–36% due to the presence of fine grinding cracks.

Materials Used in Steam Turbine Rotor Forgings. C. Sykes. (Journal of the West of Scotland Iron and Steel Institute, 1947–48, vol. 55, pp. 11–45). The types of steel used for steam turbine rotor forgings, their properties, manufacture, and inspection are described. Isothermal curves for a nickel steel (Ni 1·12%) and a low alloy steel (Cr 3·32%, Mo 0·55%) and 12 tables of properties are presented.—R. A. R.

Grain Growth in Silicon–Chromium Valve Steel. H. Allsop and P. W. Bygate. (Journal of The Iron and Steel Institute, 1949, vol. 161, Apr., pp. 318–325). The occurrence of severe grain growth in silicon–chromium valve steels, conforming in composition to B.S. En 52, and to which attention has been previously drawn, has been investigated.

It is shown that grain growth of the order noted occurs as a result of reheating critically strained material to sub-critical temperatures. If the material is in the ferritic condition at the time of straining, the degree of coarsening is influenced by the structure, the most severe growth resulting when the steel is strained in the hardened and tempered state. The temperature of the material when strained is not important, provided the ferritic condition is maintained. Alternatively, grain growth of a similar order may result from strain during hot forging if, at the time of straining, the temperature does not exceed approximately 750° C., in which case the material may be austenitic, or have transformed to widely different products. The severity of the grain growth resulting from either condition of straining is markedly influenced by the hardening temperature in the case of material strained in the hardened and tempered condition, or the initial temperature in the case of material strained during hot forging. High hardening or forging temperatures notably reduce the tendency to coarsen on reheating the strained material. Coarsened material is notch-brittle at elevated temperatures.

The tendency to coarsen increased with increasing silicon content, within the limits of B.S. En 52. The temperature necessary to effect refinement of coarsened material also increases with increasing silicon and chromium.

High Temperature Bolting Materials. E. L. Robinson. (Proceedings of the American Society for Testing Materials, 1948, vol. 48, pp. 214–238). *See* Journ. I. and S.I., 1948, vol. 160, Oct., p. 233.

Spark Tests on Steel. E. Hermelin. (Instruments and Measurements Conference, Stockholm, 1947, Transactions, pp. 97–98). [In Swedish]. The grinding sparks obtained from steels of different carbon contents and containing various alloying elements are described.—R. A. R.

METALLOGRAPHY

Symposium on the Mosaic Structure of Metals. C. Crussard and A. Guinier. (Revue de Métallurgie, Mémoires, 1949, vol. 46, Feb., pp. 61–71). The authors present a review of the evidence (X-ray, micrographic, mechanical properties, etc.) put forward in support of the existence in metals of

structural defects or of a sub-structure known as " mosaic structure." They do not attempt to define its real nature. A. E. C.

Electronic Radiography and Its Applications in Metallurgy. J. J. Trillat. (Société Française de Métallurgie, Oct. 18, 1948 : Revue de Métallurgie, Mémoires, 1949, vol. 46, Feb., pp. 79–83). The author briefly describes the principles of electronic radiography by reflection and by transmission, and gives examples of the application of the former to the examination of metallic surfaces, alloys, ores, cast iron, etc. A. E. C.

A Research and Test Method : Grinding in a Rarified Atmosphere, Associated with the Use of Electron Diffraction. R. Courtel. (Société Française de Métallurgie, Oct. 21, 1948 : Revue de Métallurgie, Mémoires, 1949, vol. 46, Jan., pp. 24–26). In order to determine the condition of a metallic surface after mechanical treatment (polishing, grinding, etc.) recourse is often had to examination by electron diffraction under grazing incidence ; sometimes clear diffraction patterns are obtained from which conclusions can be drawn, but more often either no pattern is obtained or else the large rings or haloes registered cannot be interpreted with certainty. The present author has introduced a milling appliance into the cathode vacuum chambers ; thus oxidation and contamination of the surface during milling and before examination, are prevented and the period between milling and examination is reduced. Typical diffraction patterns are shown and discussed. By making a series of tests under diminishing vacuum the author showed that the unsatisfactory patterns often obtained in air are due to oxidation of the surface.—A. E. C.

Mounting Metallographic Specimens at Room Temperature. H. S. Kalish. (Iron Age, 1949, vol. 163, Mar. 24, p. 109). The use of styrene polymerized with benzoyl peroxide for mounting soft, or low-melting-point metals and alloys for metallographic examination is described. Polymerization takes from 36 to 96 hr. according to the strength of the catalyst employed.—J. P. S.

Rapid Polish with Diamond Hand Hone. L. P. Tarasov and C. O. Lundberg. (Metal Progress, 1949, vol. 55, Feb., pp. 183–184). An account is given of the use of a vitrified bonded diamond hand hone to expedite the intermediate stages of polishing metallographic specimens. The hone can also be used for rapidly polishing high-speed steels of the high-carbon high-vanadium type, cemented carbides, and very hard non-metallic materials.—J. C. R.

The Structure of Surfaces Mechanically and Electrolytically Polished. H. Raether. (Métaux et Corrosion, 1947, vol. 22, Jan., pp. 2–17). The author presents comparative results obtained from an investigation of the structure and micro-geometry of surfaces, mechanically and electrolytically polished, using electron diffraction and electron microscope methods.—J. C. R.

The Contribution of the Electron Diffraction Method to the Study of the Beilby Layer. R. Courtel. (Métaux et Corrosion, 1947, vol. 22, Oct., pp. 157–167). A review is presented of the work of various investigators who have used the electron diffraction method to examine the nature and thickness of polished layers of metals and non-metals. A bibliography of over 60 references is appended.—J. C. R.

Metallographic Study of Solid Alloys of the Metal-Ceramic Type. N. M. Zarubin. (Zavodskaya Laboratoriya, 1948, vol. 14, Dec., pp. 1434–1436). [In Russian]. An account is given of an investigation of the action of some selective etching agents on sintered alloys of the carbides of tungsten, molybdenum, and tantalum, respectively, with cobalt. Potassium ferricyanide solution was found to be a good agent for the identification of the carbides and to have no action on the solid solutions. Whilst the reagent for the latter was found to be *aqua regia*, a mixture of nitric and acetic acids revealed their structure in the greatest detail. For the study of the structure of solid solutions in tantalum–carbide/cobalt alloys containing $< 50\%$ of carbide the reagent recommended is an ammoniacal solution of cuprous chloride. Results obtained are illustrated by photomicrographs.—S. K.

Electrolytic Polishing of Metallographic Sections. A. I. Gershevich, T. A. Mikhailova, and D. O. Slavin. (Zavodskaya Laboratoriya, 1948, vol. 14, Dec., pp. 1437–1438). [In Russian]. An apparatus for the electrolytic polishing of metallographic sections is described in which a current density of < 800 amp./sq. dm. is obtained with 6 to 12 V. The electrolyte used consists of 60–65 wt.-% phosphoric acid (sp. gr. 1·50–1·55), 15–20 wt.-% sulphuric acid (sp. gr. 1·84), and 0·5, 7·0, and 13·0 wt.-% of chromium trioxide, glycerine, and water, respectively, and contains 5 g./l. ferrous sulphate. The cell consists of the section-anode resting on a piece of glass cloth which covers the cathode. The electrolyte is gravity-fed through perforations in the latter, keeping the glass cloth thoroughly soaked and then flowing off into a trough. Photomicrographs illustrate results obtained in this way, which compare favourably with those obtained by the usual methods.—S. K.

Comparison of the Results Obtained by Various Standard Methods for the Determination of the Grain Sizes of Austenite in Steel. N. Hajtó. (Bányászati es Kohászati Lapok, 1949, vol. 82, Jan., pp. 13–15). [In Hungarian]. A series of tests was carried out to find the differences in the results obtained by the various methods of determination of the grain-size of austenite. One open-hearth and three electric-furnace steels were used. Two were carbon steels, a third contained 1% of chromium, and the fourth steel 0·55% of chromium and 1·44% of nickel. To ensure that the austenite grain-sizes were equal, the specimens were held 30 min. at 800°, 850°, 900°, 1000°, 1100°, and 1150° C. Four specimens were used for each type of steel, and different methods were applied to them to make the austenite grain-size visible. The author found that practically identical results were obtained by normalizing, arrested quench, gradient quench, and martensite etching. Fracture tests showed somewhat finer grain-sizes. The McQuaid-Ehn test, however, always showed coarser grain than that determined by other methods, and this was obviously due to the long duration of the carburization.

The results of these tests do not confirm the opinion of Hawkes who says that determination of the grain-size from the rupture surface gives unreliable results. Practically identical results were obtained for specimens of rolled steel before and after homogenization was applied.—E. G.

Retained Austenite. M. Cohen. (American Society for Metals, 23rd Campbell Memorial Lecture : Metal Progress, 1948, vol. 54, Dec., pp. 823–826). In the full lecture a comprehensive analysis is given of the theory and data concerning retained austenite in steel. In the abridged paper in *Metal Progress* the main facts about retained austenite are given, but the theory is not included.—J. C. R.

CORROSION

Some European Researches on Passivity. U. R. Evans. (Corrosion, 1948, vol. 4, Nov., pp. 545–556). A résumé is presented of the results of corrosion and passivity experiments by the author and several other European investigators. There are 27 references.—R. A. R.

Present Day Tendencies in Corrosion Research. P. Lacombe. (Métaux et Corrosion, 1947, vol. 22, Nov.–Dec., pp. 181–184). A review is presented of the recent work of various corrosion investigators.—J. C. R.

Theoretical Aspects of Corrosion and Oxidation. U. R. Evans. (Métaux et Corrosion, 1947, vol. 22, Nov.–Dec., pp. 184–191). A review is presented of some theoretical aspects of corrosion and oxidation under the following headings: Mechanism of oxidation and tarnishing without the presence of cracks ; oxidation and atmospheric corrosion in conditions favourable to cracks ; corrosion of a metal immersed in a liquid ; classification of corrosion processes according to cathodic reaction ; prevention of corrosion by soluble inhibitors ; corrosion prevention by the use of paint. A note on Wagner equations and a bibliography of 29 references are appended.—J. C. R.

What Causes Localised Corrosion ? R. B. Mears. (Steel, 1949, vol. 124, Mar. 14, pp. 86–89, 128). Basic causes of localized corrosion as, for example, at scratches or near welds, may be electrochemical. Potential differences may be set up by differential heat-treatments causing localized precipitation of constituents, as in welding, by surface roughness which causes any film formed to be less continuous than that on a smooth surface, by differential strain where plastically deformed areas may have different solution potentials from non-strained areas, by oxygen screening, or by galvanic contact with other metals.—J. P. S.

Application of Electrochemical Methods to the Measurement of the Intensity and Nature of Corrosion in Neutral Media. L. Cavallaro and A. Indelli. (Atti Accad. Sci. Ferrara, 1946–47, vol. 24, No. 2, pp. 199–211).

New Method for the Study of the Electrochemical Mechanism of Corrosion. L. Cavallaro and A. Indelli. (Atti Accad. Sci. Ferrara, 1946–47, vol. 24, No. 2, pp. 269–287).

Testing the Corrosion of Wire Specimens under the Simultaneous Application of a Constant Stress by Tension. N. D. Tomashov and V. A. Titov. (Zavodskaya Laboratoriya, 1949, vol. 15, Jan., pp. 48–53). [In Russian]. An apparatus is described in which the simultaneous corrosion-testing of 18 wire specimens could be carried out under load, the time of failure being recorded automatically. Results of tests on a steel of the following composition are given : C 0·45%, Mn 0·57%, Si 0·23%, S 0·026% and P 0·023%. The wire was tempered and patented at 800° C. and specimens, carefully degreased, were immersed under loads of 0·6, 40, 80, 100 and 120 kg., respectively, in water containing small quantities of free sulphuric acid, the sulphates of aluminium and magnesium, and the chlorides of ferric iron and calcium. The ohmic resistance of the specimens was determined after corrosion. The results of the tests are shown graphically. S. K.

Protection of Boiler Plant from Standing Corrosion. A. P. Mamet. (Elektricheskie Stantsii, 1949, No. 5, pp. 27–29 : Electrical Engineering Abstracts, 1949, vol. 52, Feb., p. 31). The effects of oxygen dissolved in water, of water vapour in empty and partially empty boilers, and of leakage of superheated steam are discussed. Distinction is drawn between the effects of shut-down periods of a few hours and those of several days. Recommendations are given for partial heating in the former case and specific details for preservation by wet and dry methods in the latter.

Materials in Boiler Feed Pump Construction. H. L. Ross. (Combustion, 1947, vol. 18, Apr., pp. 43–44 : [Abstract] Centre National de la Recherche Scientifique Bulletin Analytique, 1948, vol. 9, No. 11, p. 3186). The author discusses the corrosion and erosion of carbon steels used in the construction of boiler feed pumps, the influence of feedwater having a weak pH, and the use of special steels.

Internal Corrosion of Furnace Tubes of High-Pressure Boilers. R. C. Corey. (Combustion, 1947, vol. 18, Mar., pp. 39–42).

Rid Water of Dissolved Gas to Stop Corrosion. R. C. Ulmer. (Power, 1948, vol. 92, Aug., pp. 491–492 : [Abstract] British Coal Utilisation Research Association, 1949, vol. 13, Jan., p. 19). This is a note on the methods employed to remove air, carbon dioxide, ammonia, and hydrogen sulphide from boiler feed water.

Rehabilitation of Existing Pipe Lines. A. B. Anderson. (Yearly Proceedings of the Association of Iron and Steel Engineers, 1948, pp. 659–661). *See* Journ. I. and S.I., 1948, vol. 160, Nov., p. 342.

Corrosion and Protection of Steel Water Mains. J. H. Smith. (Journal of the British Water Works Association, 1947, vol. 29, June, pp. 173–180 : [Abstract] Centre National de la Recherche Scientifique, Bulletin Analytique, 1948, vol. 9, No. 10, p. 2392). The following points are considered : The corrosivity of drinking water ; influence of carbon dioxide content ; galvanic and chemical corrosion ; action of bacteria ; internal and external protective coatings ; protection of joints ; precautions to be taken during transport and installation of the mains ; and cathodic protection.

Underground Rust Prevention. E. Palmær. (Farg och Fernissa, 1947, vol. 11, pp. 105–108 : British Abstracts, B.I, 1948, Nov., col. 638). The effects of the underground corrosion of iron are reviewed. Methods for its prevention include the use of bituminous coatings, cement casing, coating with zinc, and cathodic protection.

Uses of Clad Steel to Combat Corrosion. M. J. Conway. (Proceedings at First Annual Meeting of the National Association of Corrosion Engineers, Apr., 1944, pp. 167–178). Having very briefly described the manufacture of steel plates clad with nickel, Monel, and Inconel, the author discussed their corrosion-resisting properties and applications in the petroleum industry.—R. A. R.

Protecting Malleable Iron Insulator Caps from Rust. F. Roll. (Archiv für Metallkunde, 1948, vol. 2, No. 8, pp. 254–256). Tests on methods of rust-proofing malleable-iron caps on porcelain insulators are described. Electrodeposited cadmium, aluminium, and zinc, as well as linseed oil were not satisfactory ; a hot-dip zinc coating was the best.—R. A. R.

Protecting Idle Machinery and Equipment against Corrosion. B. Scheifele. (Archiv für Metallkunde, 1948, vol. 2, No. 9, pp. 322–323). Methods of protecting idle machinery and plant against corrosion are briefly reviewed.—R. A. R.

Outdoor Laboratory Provides Information on Effects of Marine Corrosion. (Civil Engineering (U.S.A.), 1947, vol. 17, Oct., pp. 32–34, 78 : [Abstract] Centre National de la Recherche Scientifique, Bulletin Analytique, 1948, vol. 9, No. 11, p. 3185). A description is given of the maritime test station at Kure Beach for studying the effects of sea water, air, and marine organisms on metallic and organic coatings.

Corrosion Problems in the Manufacture of Soda Ash by the Ammonia Soda Process. G. Heinemann. (Corrosion, 1948, vol. 4, Nov., pp. 516–528). Cast iron pipes have been found satisfactory for carrying brine to a soda ash plant provided that the joints are properly made and do not leak. To protect steel storage tanks for brine the inside surface at the water line must be coated : a coating of heavy oil has been found satisfactory. To protect equipment in contact with ammoniated brine at 60–65° C., small controlled amounts of sulphide are introduced into the brine ; this tends to form a film of iron sulphide on the cast iron which resists attack. Corrosion by crude ammoniacal liquor has not yet been adequately dealt with. Coolers for hot gases from the ammonia stills are made with a cast-iron shell and tube plate and aluminium tubes.—R. A. R.

A Simple Test Method for Evaluating Corrosion Inhibitors. J. W. Ryznar and J. Greene. (Corrosion, 1948, vol. 4, Nov., pp. 505–515). A laboratory test for evaluating corrosion inhibitors is described. A solid cylindrical specimen $\frac{1}{4}$ in. in dia. × 3 in. long in an insulated holder is rotated for 24 hr. in a beaker of water to which the inhibitor is added. Means of measuring the potential of the specimen are provided. Visual examination of the specimen at the conclusion of the test is the main criterion used to evaluate the results.—R. A. R.

Mechanism of Inhibitors of Corrosion. H. H. Uhlig. (Métaux et Corrosion, 1947, vol. 22, Nov.–Dec., pp. 204–210). The author discusses the mechanism of the action of inhibitors and passivators, and the rôle of dissolved oxygen as a corrosion inhibitor. A bibliography of 33 references is appended.—J. C. R.

The Cathodic Protection of Iron by Magnesium Anodes. B. J. C. Raclot. (Métaux et Corrosion, 1947, vol. 22, Feb., pp. 28–30). The author outlines the advantages of magnesium anodes over zinc anodes for the galvanic protection of iron structures and underground piping. He then discusses the effect of pH of the surrounding media on the magnesium anodes and the importance of the constitution and positioning of the anodes.—J. C. R.

Passivation of Stainless Steel. L. Guitton. (Métaux et Corrosion, 1947, vol. 22, Apr., pp. 47–60 ; May, pp. 81–89). After reviewing the work of previous investigators on the passivity and activity of stainless steel, the author discusses passivity conditions in the presence of sulphuric acid and describes experiments to impart a stable passivation to an 18% chromium, 8% nickel, 3% molybdenum stainless steel and an 18% chromium, 10% manganese stainless steel. An electric method was tried first and excellent results were obtained using as electrolyte 20–40% by volume sulphuric acid solutions. On immersion, the steel required anodic protection but the current was lowered progressively towards zero and a stable passivity was obtained. A chemical method involving prolonged treatment in dilute nitric acid also caused passivation, but if the steel had previously been sand-blasted, after a certain time there was a sudden decrease in passivity and rapid corrosion occurred. Reference is made to the nature of sensitivity of stainless steels. It is produced by a series of treatments in acid media, the last being a nitro-fluoric solution with 10% nitric acid and 2% sodium fluoride. Finally the author gives results of experimental tests which confirm the possibility of passivation of stainless steel parts subject to hot and cold work.—J. C. R.

Action of Organic Acids on Stainless Steel. C. F. Poe and E. M. Van Vleet. (Industrial and Engineering Chemistry, 1949, vol. 41, Jan., pp. 208–210). The corrosive effects of a few common organic acids, such as are found in foods, on the stainless steel, as used for making cooking utensils, was investigated. Strips of 18/8 stainless steel resisted the attack of most dilute organic acids at 25° C. and at boiling temperatures, with the exception of boiling oxalic acid in normal concentration. The most rapid rate of attack was in the first 24 hr. Raising the temperature to boiling point usually decreased the rate of attack. From the small amounts of

chromium and nickel dissolved, contamination of foods by stainless steel utensils is unlikely.—R. A. R.

Mechanism of the Corrosion of Copper Steels. N. D. Tomashov, G. P. Sinel'shchikova and M. A. Vedeneeva. (Comptes Rendus (Doklady) de l'Académie des Sciences, U.R.S.S., 1948, vol. 62, pp. 105–108: Chemical Abstracts, 1949, vol. 43, Jan. 10, col. 98). The interpretation of the special corrosion resistance of copper steel, particularly in the presence of aeration or of oxidants, and in the absence of chlorine ions, by a passivation of iron by local currents in which the copper, bared by beginning corrosion, acts as cathode, is corroborated on a model made up of an iron–copper couple in $0{\cdot}01\,N$ Na_2SO_4 + different amounts of H_2O_2. While with an addition of 0·005 and 0·01 N H_2O_2, the current flowing from the cell is increased, further increase of H_2O_2 to $0{\cdot}1\,N$ reduces it, and with $0{\cdot}5\,N$ H_2O_2 the current drops below its value in pure Na_2SO_4 solution. Thus, while H_2O_2 in small amounts acts as a cathodic polarizer, a high amount acts predominantly as an anodic passivator. Corrosion of copper-free steel (C 0·08%, Mn 0·28%, Si 0·14%, S 0·018%, P 0·018%, Cu 0·1%) increased steadily with increasing amount of H_2O_2 in the solution; in contrast thereto copper steel (C 0·07%, Mn 0·21%, Si 0·065%, S 0·020%, P 0·022%, Cu 0·84%), under the same conditions, corroded increasingly faster with increasing H_2O_2 content only up to $0{\cdot}01\,N$, but the rate of corrosion fell rapidly with further increasing H_2O_2, until, with $0{\cdot}5\,N$ H_2O_2, the rate of corrosion was only one-tenth that of the copper-free steel. Similarly, the copper-free steel in short electric contact with an equal surface of copper first corroded increasingly faster with increasing H_2O_2, then ever slower until, with $0{\cdot}5\,N$ H_2O_2, its corrosion became slower than in distilled water. In the iron–copper model, the reversal of the trend of the rate of corrosion occurs at lower H_2O_2 contents than in the case of copper steel. This is understandable by the much more intimate electric contact between iron and copper in the copper steel, as compared with the model; the intimate contact makes for lower ohmic resistance and correspondingly higher passivating current density.

ANALYSIS

Ferrous Metallurgy. H. F. Beeghly. (Analytical Chemistry, 1949, vol. 21, Feb. pp. 241–246). A review is given of papers covering the development in the past few years of the use, in ferrous metallurgical analysis, of emission and absorption spectroscopy, X-ray diffraction, and potentiometry. The papers on new methods, especially chemical separations, are also listed; methods of separating aluminium, arsenic, boron, carbon, cerium, chromium, cobalt, columbium, tantalum, copper, germanium, hydrogen, lead, manganese, molybdenum, nickel, nitrogen, osmium, oxygen, phosphorus, silicon, sulphur, tin, titanium, tungsten, vanadium, zinc, and zirconium from iron, are briefly described and 194 references given.—J. P. S.

Chemical Analysis by Emission Spectrography. E. Lœuille. (Métaux et Corrosion, 1947, vol. 22, Mar., pp. 38–45; Apr., pp. 61–67). The present position of spectrum analysis in France is outlined and developments elsewhere in this technique are reviewed. The importance of spectrographic methods in the laboratory is emphasized and notes are given on the apparatus used. A note on the organization of the Groupement pour l'Avancement des Méthodes d'Analyse Spectrographique des Produits Métallurgiques (G.A.M.S.) is appended.—J. C. R.

An Account of the Organisation of the "Groupement pour l'Avancement des Méthodes Spectrographiques." E. Lœuille. (Instruments and Measurements Conference, Stockholm, 1947, Transactions, pp. 47–50).

Spectral Method for the Quantitative Determination of Carbon in Steels. A. P. Oleinikov and K. I. Taganov. (Zavodskaya Laboratoriya, 1949, vol. 15, Jan., pp. 59–62). [In Russian]. The spectral determination of carbon in steels using a H.F. discharge for the excitation is described. The spectrum in the region of the line C III 2297 Å was utilized, and calibration curves are given for steel with and without nickel and for cast iron for various types of photographic plates. It was found that since the H.F. discharge excites the spectra of several alloying elements, their determination could be carried out simultaneously with that of carbon. The method described was seen to be suitable for the analysis of thin surface layers, and a modification which is easily performed enables specimens of wire as small as 0·15 mm. in. dia. to be dealt with.—S. K.

Investigation of the Conditions of the Excitation of the Spectrum in the Controllable Spark Generator. E. V. Zagoryanskaya. (Zavodskaya Laboratoriya, 1949, vol. 15, Jan., pp. 53–59). [In Russian]. Experiments are described in which changes in relative intensity of various spectral lines, electrode and spark temperatures, and voltage and current changes in the production of spectra by a condensed spark fed by a special generator were measured. The results are compared with those obtained when an ordinary generator was used. S. K.

On the Spectral Determination of Carbon. K. I. Taganov. (Zavodskaya Laboratoriya, 1949, vol. 15, Jan., p. 105.) [In Russian]. The relative merits of the A.C. arc and the condensed H.F. spark for the production of the spectrum for the determination of carbon are briefly discussed, and the photographic recording of the lines is considered.—S. K.

Spectroscope Used to Identify Stainless Steel Grades. (Yearly Proceedings of the Association of Iron and Steel Engineers, 1948, p. 751). *See* Journ. I. and S.I., 1949, vol. 161, Apr., p. 397.

Photoelectric Intensity Measurements in Spectra. C. H. Dieke. (Instruments and Measurements Conference, Stockholm, 1947, Transactions, pp. 17–37). The photoelectric method of measuring the intensity of spectra is explained and discussed with several examples of its applications. The construction and behaviour of multipler phototubes are also described and explained.—R. A. R.

Advances in the Technique of Spectrographic Analysis in Great Britain. D. M. Smith. (Instruments and Measurements Conference, Stockholm, 1947, Transactions, pp. 38–46). A summary is given of work which has been done in an endeavour to standardize spectrographic analysis techniques with special reference to work sponsored by Johnson Matthey & Co., Ltd., and the British Non-Ferrous Metals Research Association.—R. A. R.

On the Use of Spectrochemical Analysis at the Boliden Mining Company. A. Danielsson. (Instruments and Measurements Conference, Stockholm, 1947, Transactions, pp. 51–55). The Boliden Mining Co. treats copper ore and gold-bearing arsenic pyrites and produces copper, lead, bismuth, gold, silver, selenium, and arsenic; some particulars are given of spectrographic technique which has been developed for determining these metals and various impurities and for controlling processes.—R. A. R.

Ultramicroanalysis by X-ray Absorption Spectrography. A. Engström. (Instruments and Measurements Conference, Stockholm, 1947, Transactions, pp. 71–75). The theory of and a procedure for microchemical analysis by X-ray absorption developed by the authors are described and its application to the determination of chromium in an iron–chromium alloy is given as an example.—R. A. R.

Adaptation of Secondary X-Ray Analysis for Industrial Purposes. A. Kochanovska. (Instruments and Measurements Conference, Stockholm, 1947, Transactions, pp. 76–78). The disadvantages of the analysis of the secondary X-ray radiation are that the exposure time is too long and the sensitivity is too small because of the low luminosity. The author has developed a method of overcoming these disadvantages. A very wide slit (more than 2 mm.) is used for limiting the secondary X-ray beam and a crystal of rock salt, curved to a logarithimic spiral, is employed as analyser. The distances of the slit and of the photographic film from the crystal are relatively small in comparison with those of ordinary X-ray spectrographs. The primary X-ray beam falls on the specimen at an angle of about 60°, and the secondary rays pass through the slit on to the curved crystal. By correct selection of the slit-to-crystal distance and curvature of the crystal, all rays of a given wavelength in a beam with its apex at the slit strike the crystal at their Bragg angle. The focusing of the reflected rays is not perfect, but the sharpness of the lines is sufficient for industrial purposes and is shown by examples which are described.—R. A. R.

Chemical Analysis by Secondary X-Ray Images. L. von Hámos. (Instruments and Measurements Conference, Stockholm, 1947, Transactions, pp. 79–80). A method of obtaining images of secondary X-rays is described. It differs from that of Kochanovska (*see* preceding abstract) in that the crystal is in the form of a sector of a tube. The method is suitable for the study of segregation in metals.—R. A. R.

Determination of the Concentration of Niobium in Corrosion-Resistant Steel with the Help of a Steeloscope. V. I. Blinov. (Zavodskaya Laboratoriya, 1948, vol. 14, Dec., pp. 1494–1495). [In Russian]. The use of the steeloscope for the identification of niobium- and titanium-containing alloy steels is briefly described, the relative advantages of various spectral ranges for the determination of these elements being discussed.—S. K.

The Absorptiometric Determination of Manganese in Steel. (Metallurgia, 1948, vol. 39, Dec., pp. 105–110). This is a report by the Methods of Analysis Committee of the Metallurgy (General) Division of the British Iron and Steel Research Association. An account is given of research on the persulphate/silver-nitrate oxidation method and periodate oxidation methods for determining manganese in steel. It was generally agreed that the latter was the more reliable as a standard method, particularly when applied to steels of widely varying compositions. The procedures are given in appendices.—R. A. R.

On the Possibility of the Interchangeability of Apparatus for Spectrographic Analysis. G. Brucelle. (Groupement pour l'Avancement des Méthodes d'Analyse Spectrographique des Produits Métallurgiques, 10th Congress, June, 1948, pp. 47–54). A comparison of two nominally similar spectrographs, and of three spark generators of different make but similar circuits has been made with a view to testing the availability of calibration results made in one laboratory for other workers. The tests were made on selected lines in the spark spectra of silicon, manganese, nickel, chromium, molybdenum, and vanadium in alloy steel. Changing the generator changes the index point for silicon while leaving those for the other five elements unaltered. The standard deviation of repeated tests on the same plate does not vary significantly for different spectrographs, and very little for different generators. A high resistance shunt across the spark gap of one generator smooths out the sparks as seen on an oscilloscope but does not improve the uniformity of the spectra.—E. VAN S.

Present State of Spectrographic Analysis of Metals in the U.S.A. P. Coheur. (Groupement pour l'Avancement des Méthodes d'Analyse Spectrographique des Produits Métallurgiques, 10th Congress, June, 1948, pp. 37–46). This is a competent critical review based on a visit to 17 American spectrographic laboratories, about half of which were analysing steel. Notes on the techniques used, including non-photographic methods, and on the precision claimed are given in some detail.—E. VAN S.

Analysis of Steel Speeded by Improved Combustion Technique. (Industrial Heating, 1948, vol. 15, Dec., pp. 2116–2118). Improvements to the combustion method of determining carbon in steel are described. Electrically heated furnaces with non-metallic resistors are used to bring the sample to 1100° C. which is about 60° higher than usual.—R. A. R.

An Electronic Carbon Analyzer for Steel Samples. A. C. Chamberlin and E. J. Serfass. (American Institute of Mining and Metallurgical Engineers, Proceedings of the 31st National Open Hearth Conference, 1948, vol. 31, pp. 266–278). A detailed description is given of the design, construction, operation, and maintenance of a new electronically operated carbon analyser. The instrument measures values proportional to the incremental alternating-current permeability of cast test specimens.—R. A. R.

Nephelometric Sulphur Determinations in Iron and Steel. G. Geuer. (Schweizer Archiv, 1948, vol. 14, Sept., pp. 275–278). A 1-g. sample of the steel is burnt with dry purified oxygen and the sulphur-bearing gases evolved are passed into a beaker and treated with lead acetate solution which, under the proper conditions of concentration and *p*H value, produces a suspension of $PbSO_3$ the intensity of which is measured optically.—R. A. R.

The Hydrogen Content of Liquid Steel Produced in Different Ways. H. Wentrup, H. Fucke, and O. Reif. (Stahl und Eisen, 1949, vol. 69, Feb. 17, pp. 117–122). The vacuum method of determining the hydrogen in molten steel is described. Results show that acid open-hearth steel has the lowest hydrogen content with an average of 6·5 c.c./100 g., whilst basic open-hearth and electric-arc furnace steels contain 9·2 and 9·4 c.c./100 g. respectively. Causes of the differences are discussed. Differences in hydrogen content at tapping and in the ingot mould, and the effect of mould coatings are pointed out. The relation between hydrogen contents and "flakes" is also dealt with.—R. A. R.

The Determination of Tungsten and Niobium in Steel. L. M. Budanova and K. D. Gavrilova. (Zavodskaya Laboratoriya, 1949, vol. 15, Jan., pp. 7–11). [In Russian]. A satisfactory method for the determination of tungsten in niobium-containing steels is described in which 1–2 g. of the sample are dissolved in a mixture of sulphuric and phosphoric acids, and, after iron, nickel, cobalt, chromium, etc., have been precipitated as hydroxides, the tungsten is determined colorimetrically under carefully controlled *p*H conditions. Neither niobium nor molybdenum interferes with this determination. The separation of tungsten and niobium by cupferron in the presence of fluorides was investigated and forms the basis for the determination of niobium in tungsten-containing steels, this being carried out gravimetrically after the separation.
S. K.

INDUSTRIAL USES AND APPLICATIONS

Which Wrought Stainless Steel? S. P. Watkins and R. J. Berkol. (Machine Design, 1948, vol. 20, Mar., pp. 106–112; Apr., pp. 174–179). Particulars are given of the analyses, properties, and applications of the austenitic chromium–nickel and the martensitic chromium stainless steels as classified by the American Iron and Steel Institute

Use of Steel in Top Slicing. J. L. Bruce, G. W. Nicholson, and J. G. Tate. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 2458: Mining Technology, 1948, vol. 12, Sept.). An account is given of the top slicing method of mining copper pyrites on the island of Cyprus, where steel posts, collar braces, and steel matting are used instead of timber.—J. C. R.

Steel Conservation in Lightweight Freight Cars. M. Barloon. (Iron Age, 1949, vol. 163, Feb. 3., pp. 102–109). The economic advantages of lightweight construction of railway wagons are outlined; the additional cost of these vehicles is attributed less to the cost of the steels of which they are made than to the uneconomical quantities in which they are ordered. Where the entire cost is $250, only $75 is for more expensive steel, such as "Cor-ten," "N-A-X," "Hi-Steel," "Otiscoloy," etc. If full use were made of the proportion of these steels in railway wagon and other construction, from 4 to 5·6 million tons of steel could be saved per year.—J. P. S.

HISTORICAL

Ancient Furnaces for Direct Reduction of Iron from Ores of the Silurian Basin in Central Bohemia. J. Širokich. (Hutnické Listy, 1948, vol. 3, Sept., pp. 270–273). [In Czech]. The discovery of remnants of ancient direct-reduction furnaces is described and an explanation is given of the functioning and chemical processes in these furnaces which were operated at only 700°–900° C. Ores containing 45–48% of iron and 10–30% of silica were broken up into 15–25 mm. pieces which were mixed with an equal volume of charcoal. This mixture was charged into the furnace in quantities of about 70 cu. dm. at a time. Such a charge yielded 2·5–5·8 kg. of iron, and 5 to 12 charges were necessary to obtain about 30 kg. of iron, for which a time of 12 to 24 hr. was required. To withdraw the iron it was necessary to break open the lower part of the furnace wall, which was then repaired so that the operation could start afresh.—E. G.

The Steel Age. Sir William Larke. (Transactions of the Institution of Engineers-in-Charge, 1948, vol. 54, Nov., p. 75; Dec., pp. 80–93). A historical account of the development of the production of steel from the earliest use of meteoric iron up to the present time is presented and followed by some facts illustrating the far-reaching results of its application.—R. A. R.

A Brief History of Alloy Steel. C. A. Zapffe. (Metal Progress, 1948, vol. 54, Oct., pp. 459–467). A general survey is presented of developments in the metallurgy and applications of alloy steels from earliest times to the present day.—J. C. R.

How the Cutler's Art Has Guided the Modern Steel Industry. C. More. (Métallurgie, 1949, vol. 81, Feb., pp. 9, 11). The author briefly sketches the development of cutlery manufacture from the days when man first fixed a blade in a flint handle up to the present century.—J. C. R.

Sidelights on the Early History of Some Metals. F. A. Fox. (Sheet Metal Industries, 1949, vol. 26, Feb., pp. 347–351, 357). Some incidents revealing the early history of the production and use of metals are described. With regard to

iron it is stated that a storehouse containing about 160 tons of forged iron bars, most of them up to 6 in. thick and up to 20 in. long, was found in the ruins of the palace of an Assyrian king, built about 710 B.C.—R. A. R.

ECONOMICS AND STATISTICS

Plan for East Coast Steel Plant. H. A. Knight. (American Institute of Mining and Metallurgical Engineers : Journal of Metals, 1949, vol. 1, Feb., Section 1, pp. 6–9). The erection of steelplants on the east coast of U.S.A. is discussed, possible sites being on Staten Island, Norfolk, Va., and Bridgeport on Long Island Sound. Although removed from supplies of raw materials the plants would be nearer to steel consumers than are present producers. Possible sources of supply of iron ore, coal, and limestone are considered and supplies of industrial scrap are thought to be promising. Plans are outlined for the equipment of such a plant with estimates of the probable cost.—J. C. R.

Production of the Polish Metallurgical Industry and Its Export Prospects. (Bányászati és Kohászati Lapok, 1948, vol. 3, Nov. 15, pp. 301–302). [In Hungarian]. Immediately after the war great efforts were made to restart production in factories which escaped large-scale war damage and the output of the metallurgical industry in that year was 46·7% of the 1938 figure. The production improved rapidly and in 1947 the output surpassed that of 1938. The exported products of the metallurgical industries are analysed.—E. G.

Review of the Production of Foundry Coke in the Ostrava-Karvin Region (Czechoslovakia) in 1947. V. Samuel. (Hutnické Listy, 1948, vol. 3, Sept., pp. 276–279). [In Czech]. About 180,000 metric tons/annum of foundry coke are being produced in the Ostrava-Karvin region, but this represents only about 10% of the total coke consumption of that region. It is intended to increase this production to 400,000 tons/annum. The coal for this purpose must be very low in volatile matter and have good coking properties. The author believes that the ash content of the coke, is not of major importance as some of it combines with coke particles in the coke-oven to form a mixture which will burn in the cupola.—E. G.

Post-war Development of the Czechoslovak Foundry Industry. A. M. Plešinger. (Hutnické Listy, 1948, vol. 3, Nov., pp. 325–329). [In Czech]. The post-war situation of the Czechoslovak foundry industry is described, and the changes due to nationalization are explained. The production figures are compared with the appropriate pre-war figures and with those in the U.S.A. and the U.S.S.R. Productivity is about equal to the average productivity of French foundries, *i.e.*, about twelve tons per worker per annum.—E. G.

Volta Redonda Steel Works, near Rio de Janeiro, (Revista de Química Industrial, 1947, vol. 16, No. 186, pp. 21–22 : British Abstracts, B.I, 1948, Dec., col. 713). The Volta Redonda works of the Compania Siderurgica Nacional is among the largest of its kind in South America. Its annual production is : Cast iron 300,000 tons, steel 256,000 tons, semi-manufactured steel (rolling-mill, etc. products) 200,000 tons, and coke (mostly for its own use) 400,000 tons. In addition 15,200,000 litres of tar, 3,888,000 litres of benzene, and 5200 tons of ammonium sulphate are produced. The works layout has been designed to permit of quadrupling eventually. Many chemical by-products are also produced.

Do You Use Your Cost System ? R. L. Lee. (Proceedings of the Institute of British Foundrymen, 1947–48, vol. 51, pp. A190–A201). *See* Journ. I. and S.I., 1949, vol. 161, Jan., p. 79.

MISCELLANEOUS

Trends in Electrification of America Steel Industry. L. A. Umansky. (Transactions of the American Institute of Electrical Engineers, 1948, vol. 67, Part I, pp. 784–795). The developments and changes during the last twenty-five years in the American steel industry by the increased use of electric power are reviewed. The rôle of electricity in improving processes and in lowering the costs of production is pointed out.—R. A. R.

Design for Ferrous Metallurgy Laboratory. J. Rist. (Publications de l'Institut de Recherches de la Sidérurgie, 1948, Series B, No. 4). The purpose, design, and personnel required for the contemplated laboratories for the Institut de Recherches de la Sidérurgie are outlined.—R. F. F.

The Use of Phosphates for the Treatment of Water. M. Reignier. (Technique de l'Eau, 1948, Feb., pp. 9–13 : [Abstract] Revue Universelle des Mines, 1949, Series 9, vol. 5, Jan., p. 33).

Economic Factors Governing the Selection of Feedwater Treatment. F. N. Kemmer. (Combustion, 1948, vol. 20, Aug., pp. 30–34).

Prevention of Fatigue as a Factor Increasing Labour Efficiency. E. Ucka. (Hutnik, 1948, vol. 15, July–Aug., pp. 328–334). [In Polish]. Interesting points of view are brought out in the discussion of the prevention of fatigue as a factor in increasing labour efficiency. Many methods of saving the worker's time are shown.—W. J. W.

Estimating Costs of Industrial Oxygen. M. J. Conway and J. J. Hogan. (American Institute of Mining and Metallurgical Engineers, Electric Furnace Steel Conference, Dec. 4–6, 1947, Proceedings, vol. 5, pp. 67–78). Having briefly described the low pressure continuous process for producing oxygen on a large scale, the authors describe and explain a system for calculating the unit cost of making oxygen in plants of almost any size. Finally, an example is given of the calculation of the advantage of using oxygen for decarburizing in a given steelmaking operation.—R. A. R.

Theoretical and Experimental Methods for the Study of Closed Cycle Automatic Control Systems. L. von Hámos. (Instruments and Measurements Conference, Stockholm, 1947, Transactions, pp. 129–138).

The Interrelation of the Engineering and Metallurgical Industries. Sir Arthur P. M. Fleming. (Proceedings of the Institute of British Foundrymen, 1947–48, vol. 51, pp. B2–B7). *See* Journ. I. and S.I., 1948, vol. 159, June, p. 238.

Diversified Techniques Involved in Multipurpose Quality Control. O. W. Hitchcock. (Steel, 1949, vol. 124, Apr. 4, pp. 98–101, 128). Statistical methods used to control the manufacture of small aircraft parts are described ; tables indicate the size of samples to be taken from lots of varying size when either 0·5% or 1% rejects may be permitted in the out-going material. The methods are applicable to other manufactures.—J. P. S.

How Steel Producers View Steel Compositions and Specifications. C. M. Parker. (Steel, 1949, vol. 124, Mar. 14, pp. 91–98). The advantages and disadvantages, and the effects on the steelmaker, of ordering steel by (1) product description, (2) chemical composition, (3) mechanical properties, (4) hardenability limits, (5) description of the part involved and the fabricating methods available, or (6) formal specifications such as ASTM, are discussed.—J. P. S.

Steelworks Maintenance. G. S. T. Martin. (Journal of the West of Scotland Iron and Steel Institute, 1947–48, vol. 55, pp. 46–68). *See* Journ. I. and S.I., 1948, vol. 158, Mar., p. 414.

Modern Methods of Feedwater Conditioning for High Pressure Boiler Plants. S. D. Distelhorst. (Yearly Proceedings of the Association of Iron and Steel Engineers, 1948, pp. 463–467). *See* Journ. I. and S.I., 1948, vol. 160, Dec., p. 455.

Cleaning of Service Water in Steel Plants. S. P. Kinney. (Yearly Proceedings of the Association of Iron and Steel Engineers, 1948, pp. 124–127). *See* Journ. I. and S.I., 1948, vol. 159, June, p. 238.

Prevention of Mechanical Failures in Steel Plant Equipment. O. J. Horger and G. A. Stumpf. (Yearly Proceedings of the Association of Iron and Steel Engineers, 1948, pp. 337–345). *See* Journ. I. and S.I., 1948, vol. 159, Aug., p. 463.

Air Conditioning of Crane Cabs. R. D. Darrah. (Iron and Steel Engineer, 1949, vol. 26, Mar., pp. 96–108). Factors affecting the temperature and humidity of the atmosphere inside steelworks crane cabs are discussed. Systems for reducing the amount of dust, dirt, fumes, and gases and for cooling the atmosphere are described.—J. C. R.

The Organization of Industrial Research. J. W. Jenkin. (Journal of the Birmingham Metallurgical Society, 1948, vol. 28, Dec., pp. 257–275). The organization of industrial research is discussed under the following headings : Definition of terms ; the relation of research to development ; the division of functions ; staffing ; progress and planning ; and costing.

R. A. R.

Tasks of the Ferrous Metallurgist in the Province of Material Testing in War. H. Schultz. (Verein deutscher Eisenhüttenleute, 1940, Confidential Report No. 8). The consequences of raw material shortages in the German iron and steel industry are discussed and classified in three groups as shortage of iron itself, alterations in the condition of the products

consequent on changes in ores and switch-over in production processes, and shortages of alloying elements. Reduction of section coupled with higher strength, application of welding, improved design, and substitution of other materials (*e.g.*, aluminium sheet) are suggested to deal with the first difficulty. The consequences of changes in the ores as regards increased arsenic content, increased sulphur, and contamination of the steels with copper and chromium from scrap, and directives for the applications of basic-Bessemer and open-hearth steel, are discussed. In the very important question of alloying elements, the availability of silicon, vanadium, the relative shortages of other elements, the degrading and modification of steel compositions, surface impregnation with metals, and use of newly developed protective coatings are dealt with.

G. W. A.

The Training of a Metallurgist. J. H. Andrew. (Iron and Steel, 1949, vol. 22, Mar., pp. 81–83). Suggestions are made for the staffing and equipping of the metallurgical department of a university, and the subjects to be included in the metallurgical course.—J. C. R.

BOOK NOTICES

Föppl, H. "*Messung der Spannungen und der plastischen Verformungen an oberflächengedrückten Probekörpern.*" Heft. 41. Mitteilungen des Wöhler-Instituts). 8vo, pp. 67. Illustrated. Brunswick, 1948. Friedr. Vieweg und Sohn, (Price 4 DM.).

This 67-page communication from the Wöhler Institute at Brunswick, is devoted to a study of the stress and plastic strain created in cylindrical test-pieces by surface rolling. In the introduction the author points out that during the last ten years, cold working the surface of highly stressed machine parts has found wide industrial application; the parts may be hammered, rolled, or shot-peened, the preferred method depending upon the shape of the part. Dr. Föppl has confined himself to a consideration of the effect of surface-rolling on cylindrical test-pieces 200 mm. long and 44 mm. in dia. The surface-rolling was carried out in a lathe, the bars being fastened at one end in a three-jaw chuck whilst the other end was supported by the fixed centre. A special device somewhat of the nature of a roller steady was used to force the hardened steel rollers against the surface of the test-piece while the lathe was running.

The effect of roller form, number of rollers, pressure of rollers, traverse of rollers per revolution of the test bar, repeated rolling, etc., is exhaustively considered. Briefly, the test-pieces were measured before and after rolling and also after the cold-worked surface layer had been dissolved away, and the results were used to calculate stress and plastic deformation.

The communication gives all the necessary practical details of the experimental work and all the theoretical work is clearly laid out.

The author has a pleasant style and fortunately avoids the use of long words and sentences so beloved of many German authors. The communication is indeed written in such simple language that only an elementary knowledge, comparatively speaking, of technical German is necessary in order to be able to understand it.—J. Ferdinand Kayser.

Fox, R. "*Ferrous Materials for the Engineer.*" 8vo, pp. xii + 159. Illustrated. London, 1948 : Charles Griffin and Co., Ltd. (Price 12*s.* 6*d.*)

The author has written this small book to provide a helpful source of reference to the industrial engineer and to stimulate the student to refer to the more comprehensive textbooks for the intricacies of the subject. In Chapter 1 he touches upon types of fractures, microscopic examination and the interpretation of photomicrographs, sulphur prints, macro-etching, and crack detection. Chapter 2 deals with the testing of materials and outlines the commonly used tests, both mechanical and physical. It includes definitions of the principal terms used in testing. The heat-treatment of steel is dealt with in Chapter 3 and particulars are given of various methods of heat-treating. In addition, inverse cooling rates, dilatation curves and the iron/iron-carbide diagram are explained, and the control of quality during heat-treatment discussed. Chapter 4 is devoted to the ferrous materials available to the engineer and outlines the characteristics and uses of cast irons, carbon steels, and alloy steels. Useful conversion tables are appended. The book contains 14 tables and 71 illustrations.—R. E.

Garside J. E. "*Process and Physical Metallurgy.*" 8vo, pp. xix + 499. Illustrated. London, 1949 : Charles Griffin and Co., Ltd. (Price 40*s.*).

In the author's own words the object of this book is "to give an account of some of the chief aspects of Process and Physical Metallurgy. The viewpoint of the user of metals and alloys, particularly that of the engineer, has been kept in mind throughout."

The larger proportion of the book is given to a fairly reasonable outline of some of the main principles of physical metallurgy followed by consideration of the principal commercial metals and alloys with respect to their metallography, treatment, properties, and applications. There are several surprising omissions, for example, no mention of fatigue can be found by the reviewer, there is scanty reference to creep in the section on heat-resisting steels, and there is no discussion of testing, apart from a short appendix defining the characteristics of the tensile test and a table comparing the various hardness scales. The general chapters on physical metallurgy are not illustrated with photomicrographs, although a few sketches are given. However, in spite of the fact that no special paper is used for reproducing the photographs, with regard to the subsequent chapters it is clear that many of the original photomicrographs, especially of non-ferrous metals, are mediocre and a number do not justify inclusion in the book.

Process metallurgy constitutes about a third of the text and is mainly treated first. General treatment comprises short elementary accounts of the main casting and working processes, fuels, refractories, moulding sands, and in addition, of certain aspects of metal-winning. A specific criticism of the brief discussion of tube-making (p. 122) is that the reader gains the erroneous impression that the Mannesmann process yields a finished tube.

Admittedly it has only been the author's intention to touch lightly on metal-winning techniques, nevertheless, a balanced summary would have been preferable to the sketchy treatment accorded to odd aspects in the two short chapters with the titles "Ores and Ore-Dressing," and "Fluxes and Slags," and partly in the chapter on "Furnaces." Further, the contents of the first-mentioned chapter, which occupy just over two pages of text, are unsatisfactory. The preliminary definitions of mineral and ore are not inaccurate (although native elements as well as compounds are minerals), but they may be misleading. In addition to jigging, flotation, and magnetic separation, vanning is mentioned, yet tabling and heavy-media separation are omitted. Moreover, surely copper or tin would have been more appropriate examples in the following (p. 12) : "In many cases ore dressing is very important, particularly with many of the non-ferrous metals such as aluminium."

In addition to the above there are short chapters on pyrometry, iron and steel making, defects in steel ingots, and the joining of metals. There is a complete absence of references to further reading on any topic.

Undoubtedly "Process and Physical Metallurgy" offers in one volume a collection of useful information covering a wide field, and in the main it is accurate and soundly written. The book would serve satisfactorily for certain types of students, especially in engineering. However, there is nothing outstanding in the work and forty shillings is an excessive price.—Alan R. Bailey.

Corrigenda

In the March *Journal*, page 265, column 2, abstract 1, the first author's name should be S. Wernick. In the May *Journal*, page 127, second book notice, the second author's name should be A. Beckel, and on page 110, column 1, abstract 7, the first sentence should read : "The radiation of carbon dioxide and water vapour in gaseous fuels . . ."

SCHACK, A. "*Die Industrielle Wärmeübergang für Praxis und Studium mit Grundlegenden Zahlenbeispielen.*" 3. Auflage, 8vo, pp. x + 385. Illustrated. Düsseldorf, 1948 : Verlag Stahleisen m.b.H. (Price 24.50 DM.).

Apart from certain relatively minor changes, necessitated by recent progress in the field of heat transfer, the third edition of Schack's book follows the earlier editions closely in most respects. This in itself may serve as a recommendation, reflecting as it does not only the relative completeness of the existing knowledge of the processes involved in heat transfer, but the author's unqualified mastery of a very wide field of enquiry.

The book, which contains almost 400 pages, is subdivided under nine main headings.

After a brief introductory chapter in which the three kinds of heat transfer are outlined and the surface conductance is defined, the author proceeds to deal with heat transfer by conduction in composite pipes and walls, and in porous brickwork. Heat transfer by condensing vapours and by radiation from solids and from flames is also considered in detail. The most extensive subsection of this—the second—heading deals with convection by laminar and turbulent flow, from the point of view of measurement as well as theory. The subsection dealing with heat transfer from plates and tubes in static air and under turbulent conditions has undergone considerable revision, mainly because of the more reliable data on the conductivity of gases which have become available in recent years.

Under the third main heading 70 pages are devoted to recuperators and regenerators. Most of this work is entirely new, in particular the section dealing with the influence of the thermal properties of the lining and of temperature changes in the gas and air-blasts on the generator performance. Many useful formulæ are derived and clearly set out in the text.

The remaining chapters deal with the calculation of the heat-transfer in industrial furnaces and with pressure losses in pipes and bundles of pipes. Further detailed work on convection by turbulent gas streams and the effect of pressure losses on the heat-transfer efficiency is also presented. These sections of the book should be of particular interest to the engineer-designer.

The whole book has in fact been written with a constant regard to the requirements of the practical engineer and the student. The mathematical tool is used freely throughout, which, in view of the nature of the subject, is of course entirely justified. Numerous tables are appended at the end of the book, and charts of functions are included at the appropriate places in the text, so that it should be possible to solve even rather involved problems of heat transfer merely by the additional use of a slide rule or at most perhaps of log tables. Although the author very commendably seems to have succeeded in presenting the mathematics in the simplest form possible, he has nowhere evaded giving—even if in smaller print only—detailed derivations of the formulæ used.

Several pages on the early part of the book are devoted to an introduction to the Theory of Similarity. The treatment is not by any means exhaustive, but in view of the extent of this subject and the mathematical complexity inherent in its formulations, one cannot but agree with the author when he limits himself to a short exposition of this subject within the framework of the present book. Relating to the subject matter of every chapter, there are a number of worked examples at the end of the book dealing for the most part with such problems as are often to be met with in practice.

The most important formulæ in the text are enclosed in rectangular frames of heavy type, so as to make them more conspicuous, and to the student probably more easy to remember. There is a gratifying air about this method of emphasizing final results—probably because it is reminiscent of the satisfaction derived from underlining one's own experimental results in the note-books of early student days.

A table of the symbols used appears at the end of the book ; and to a reader who is using the book for reference purposes their presence may for some time remain unnoticed. In the first English edition the symbols were at the beginning, which seems more effective, though perhaps rather less appealing from the aesthetic point of view.

On the whole, the book is a fount of information, invaluable to the engineer and student, and can serve as a model to all those who write for a wide section of the readers of technical literature.—PAUL FELTHAM.

NEW PUBLICATIONS

AMERICAN SOCIETY FOR METALS. "*Sleeve Bearing Materials.*" Edited by R. W. Dayton. 8vo, pp. vi + 256. Illustrated. Cleveland, Ohio, 1949 : American Society for Metals. (Price $5.00).

BARNES, R. M. "*Motion and Time Study.*" Third Edition. 8vo, pp. 559. Illustrated. London, 1948 : Chapman and Hall, Ltd.

CHAUSSIN, C. and G. HILLY. "*Métallurgie.*" Tome II. "*Elaboration des Métaux.*" 8vo, pp. xiii. Illustrated. Paris, 1949 : Dunod. (Price 560 frs.).

COAL INDUSTRY SOCIETY AND INSTITUTE OF FUEL. "*Coal Production, Distribution, Utilisation.*" Edited by P. C. Pope for the Joint Publications Committee of the Coal Industry Society and the Institute of Fuel. Pp. 236. London : Industrial Newspapers Ltd. (Price 12*s.* 6*d*).

COHEN, A. "*Rationelle Metallanalyse.*" 8vo, pp. 416. Illustrated. Basle, 1948 : Verlag Birkhäuser.

DOBIE, W. B. and P. C. G. ISAAC. "*Electric Resistance Strain-Gauges.*" 8vo, pp. 114. Illustrated, London, 1948 : English Universities Press. (Price 15*s.*).

"*Effects of Alloying Elements and the Tensile and Hardness Properties of Carbon and Alloy Steel.*" Pp. 143. Pittsburgh, 1948 : Heppenstall Co.

FLORES, M. M. "*Acero y hornos electricos.—Fabricación, tratamentos, aplicación, pruebas.*" 8vo, pp. 463. Illustrated. Madrid, 1947. Editorial Dossat S.A. (Price 170 pesetas).

FLORES, M. M. "*Formulario Metalurgico.—Tablas y datos prácticos para uso de Fundidores, Metalúrgicos y Receptores de Materiales.*" 8vo, pp. 245. Illustrated. Madrid, 1947. Editorial Dossat S.A. (Price 70 pesetas).

GOETZEL, C. G. "*Powder Metallurgy.*" Vol. 1, pp. 806. New York : Interscience Publishers Inc. (Price $15.00).

HANCOCK, L. H. "*Engineering Materials and Processes.*" London : Sir Isaac Pitman and Sons, Ltd. (Price 15*s.*).

HEMINGWAY, E. L. "*Wear and Surface Finish.*" 4to, pp. 80. Illustrated. Madison, Wis., 1947 : Gisholt Machine Co.

JUSTI, E. "*Leitfähigkeit und Leitungsmechanismus fester Stoffe.*" Pp. xii + 348. Göttingen, 1948 : Vandenhoek und Ruprecht. (Price 15 DM.).

KACHANOW, L. N. and others. "*Plastic Deformation, Principles and Theories.*" 8vo, pp. 192. Illustrated. Brooklyn, N.Y., 1948 : Mapleton House. (Price $8).

KEENAN, J. H., and J. KAYE. "*Gas Tables.*" 8vo, pp. 236. New York : John Wiley and Sons, Inc. (Price $5).

PADDA, L. "*Forni industriali a combustione.*" 8vo, pp. 508. Illustrated. Milan, 1948 : E. Hoepli.

PATTY, F. A. "*Industrial Hygiene and Toxicology.*" Vol. 1, pp. 531, vol. 2, pp. 750, New York : Interscience Publishers.

RODÉS, R. C. "*Metales y Aleaciones.—Su constitución, estructura, propiedades y tratamentos.*" Tomo I.—"*Constitución y estructura.*" Madrid, 1948. Publicaciones del Instituto Nacional de Técnica Aeronáutica.

SCHWARTZ, A. M. and J. W. PERRY. "*Surface Active Agents. Their Chemistry and Technology.*" Pp. xi + 579. New York : Interscience Publishers. (Price $10).

"*Thorpe's Dictionary of Applied Chemistry.*" Fourth Edition. Vol. IX. "*Oils, Fatty to Pituitary Body.*" London : Longmans, Green and Co. (Price 80*s.*).

WALKER, R. C. "*Photo-Electric Cells in Industry.*" London : Sir Isaac Pitman and Sons, Ltd. (Price 40*s.*).

WILSON, A. J. C. "*X-Ray Optics : The Diffraction of X-Rays by Finite and Imperfect Crystals.*" La. 8vo, pp. vii + 127. Illustrated. London, 1949 : Methuen and Co. Ltd. (Price 6*s.*).

WYCKOFF, R. W. G. "*Crystal Structures.*" Section 1. (Loose-Leaf). 8vo, pp. viii + 442. New York and London, 1948 : Interscience Publishers Inc. (Price 48*s.*).